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TRANSACTIONS

of the American Society for Steel Treating

CONTENTS

Steels Made Without Manganese
—S. B. Ritchie 193

Tool Steel From the Consumer's Standpoint
—H. G. Keshian 233

Comparative Tests of Hardened Steel by Means of Penetration Testing Apparatus
—N. N. Sawin and E. Stachrowski 249

Thermomagnetic Phenomena in Steel
—R. L. Sanford and G. A. Ellinger 265

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RAY T. BAYLESS, *Editor*

Vol. XX September, 1932 No. 3

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STEELS MADE WITHOUT MANGANESE

By S. B. Ritchie

Abstract

The manufacture and tests of steel made without manganese in high frequency induction furnaces are described. Zirconium was used to replace manganese. Macrographs and photomicrographs are shown to illustrate structure of centrifugal castings and forgings.

It appears that by special treatment steel that will not be red-short and that will meet the strength requirements for cannon can be made with little or no manganese.

MANGANESE is classed by the War Department as a strategic material; that is, a material that is not available in this country in suitable quantity and form to supply the needs of the steel industry in a national emergency were the supply from abroad shut off, and for which there is no adequate substitute. Work was therefore undertaken at the Watertown Arsenal (under the stimulating interest of Gen. T. C. Dickson, commanding officer, and with the cooperation of the laboratory staff) to develop a satisfactory substitute for all or the greater part of manganese now used in making steel. The program so far has been directed to the making of steel suitable for cannon without using any manganese, on the assumption that this would be the worst condition ever to be met. It is realized that the data given are neither complete nor conclusive. This paper is in the nature of a preliminary report, and further work is necessary to prove that steel made without manganese even though it may pass the minimum gun-forging requirements (transverse) of 60,000 pounds per square inch proportional limit, 90,000 pounds per square inch ultimate strength, 18 per cent elongation, 30 per cent reduction of area, and a tensile-charpy of 24 foot-pounds is suitable for cannon.

The Manganese Supply

There is available in this country a large quantity of manganese

A paper presented before the Thirteenth Annual Convention of the society held in Boston, September 21 to 25, 1931. The author, Captain S. B. Ritchie, a member of the society, was stationed at the Watertown Arsenal during the time this research work was done. He is at present stationed at Fort Sam Houston, San Antonio, Texas. Manuscript received July 1, 1931.

ores. These ores, however, are mostly lean and it would require considerable time to obtain ferromanganese in quantity suitable for steel making. Domestic production of manganese ore is below the normal peace-time requirements, and in an emergency should we be forced to depend upon our own resources, the time lost even to develop the limited supply of native manganese might be fatal. In this connection the United States Bureau of Mines has been working for a solution of the emergency manganese problem. Dr. Herty shows[1] that by using special manganese-silicon alloys for the deoxidation of steel, in the form of high-silicon spiegel (20 per cent manganese, 4 per cent silicon, and 4.25 per cent carbon), for basic heats, or a special alloy (30 per cent manganese, 5.5 per cent silicon, and 2.5 per cent carbon), for acid heats, a great reduction in ferromanganese requirements is effected, and in the grades of steel which constitute our largest tonnage, ferromanganese can be eliminated altogether by the substitution of ordinary spiegel. He states that there is a tremendous tonnage of low grade manganese ore in this country which is suitable for spiegel production, either as high-silicon spiegel or ordinary spiegel. But there are only a few deposits and a small tonnage of the ferro-grade of ores available from which ferromanganese (80 per cent grade) can be manufactured.

Zirconium, A Possible Substitute For Manganese

During the World War it was rumored that the Germans were using zirconium-bearing steels for shells and for armor plate, and it was even claimed that the Big Berthas were lined with zirconium steels. Much interest was aroused in America in regard to this supposedly superior alloy metal. The Ford Motor Company during the war made light armor plate containing zirconium. This material had excellent ballistic properties. Duplication, however, apparently was uncertain. Others in the Government service took up the investigation at that time. The Bureau of Standards, the Bureau of Mines, and the Electro Metallurgical Company all did extensive research work to ascertain the properties of this metal as an alloying element and as a deoxidizer. Some reports on zirconium were favorable, others unfavorable. The Electro Metallurgical Laboratory under the direction of Dr. F. M. Becket continued the work suffi-

[1]C. H. Herty, Jr., "Report of Investigations No. 330, U. S. Bureau of Mines, January, 1931."

ciently to show that zirconium was not only a powerful deoxidizer, but also could replace manganese in steel for certain purposes. In 1923 and 1924, Feild[2] discussed the deoxidizing and desulphurizing qualities of zirconium and pointed out the beneficial effects to be derived from zirconium in steel. He noted "that the reaction between zirconium and sulphur in molten steel may be represented by the simple equation, $Zr + 2S = ZrS_2$, and that the zirconium sulphide thus formed occurs in the finished steel in the form of gray-colored inclusions, closely resembling manganese sulphide and, like the latter, exhibiting plasticity at rolling temperatures." In some of the steels described in the present paper the writer has observed small dove-colored inclusions believed to be zirconium sulphide.

Since Feild's admirable work, others have made forging steels containing little or no manganese by using zirconium to eliminate red-shortness. The classic example of steel made without manganese, and without zirconium, is Armco iron, but that material has a particular range within which it has to be rolled and at certain temperatures it cracks badly in forging or rolling.

Manufacture Of Steels For Test

The steels, the tests of which are reported herein, were all made in high frequency induction furnaces, the same as used for heats for casting guns by the centrifugal process. It is realized that this type of furnace has not yet come into general use as a melting unit for tonnage production, and that the open-hearth furnace must be depended upon to furnish the great quantity of steel needed in a national emergency. In view of the fact, however, that the centrifugal process has been established as a method for making cannon, steel even in limited quantity made in induction furnaces may be of inestimable value since they are admirably adapted to melt steel for casting guns centrifugally. A total of seven heats were made, four for ingots and three for centrifugal castings. Table I shows the number of the heat, the weight, size and type of ingot, or casting, and the composition of each.

The steel for the 375-pound ingots was melted in a 500-pound

[2]A. L. Feild, "Some Effects of Zirconium in Steel," *Transactions*, American Institute of Mining and Metallurgical Engineers, 1923, vol. 69, p. 848.

A. L. Feild, "Effect of Zirconium on Hot-Rolling Properties of High-Sulfur Steels and the Occurrence of Zirconium Sulfide," *Transactions*, American Institute of Mining and Metallurgical Engineers, 1924, vol. 70, p. 201.

Table I

Composition of Steels Tested

Heat No.	Description	Carbon	Manganese	Silicon	Volumetric Sulphur	Phosphorus	Molybdenum	Vanadium	Zirconium
173	375 lbs. 6½"x6½" square	0.365	0.07	0.31	0.017	0.009	0.25	0.16	0
239	375 lbs. 6½"x6½" square	0.30	0.06	0.24	*0.016	0.015	0.30	0.12	0.017
249	375 lbs. 6½"x6½" square	0.29	0.05	0.225	**0.015	0.017	0.33	0.105	0.0238
263	5000 lbs. 21-inch fluted	0.335	0.06	0.29	0.014	0.016	0.30	0.14	0.0496
C-207	690 lbs. centrifugal casting for 75 mm. Pack Howitzer	0.35	0.06	0.24	0.014	0.018	0.27	0.10	0.0204
C-209	690 lbs. centrifugal casting for 75 mm. Pack Howitzer	0.34	0.065	0.23	0.015	0.013	0.32	0.14	0.0410
C-218	690 lbs. centrifugal casting for 75 mm. Pack Howitzer	0.36	0.08	0.210	0.019	0.013	0.29	0.16	0.0265

*Gravimetric Sulphur = 0.023. } Ratio indicated assumed for other heats.
**Gravimetric Sulphur = 0.024. }

induction furnace; for the centrifugal castings in a 1000-pound induction furnace, and for the 5000-pound ingot in a 2¼-ton induction furnace. For melting in these furnaces generally all the charge is added at the start except a small amount of final deoxidizer, which is added just before the pour. In these heats about one-half the zirconium addition was made at the start and the balance five to ten minutes before pouring. In order to reduce as far as possible any inclusion of manganese, Armco iron was used as the base charge, with wash metal and ferro-alloys to give the desired composition. The furnace was sealed with a cover to exclude air during the melting period. Table II shows the additions to the charge of carbon, silicon, zirconium, vanadium and molybdenum with the recovery in each case.

The recovery of zirconium was much less than might be expected. However, the Armco iron used in the charge was high in total oxygen content (0.086 per cent). No slag other than that formed from the furnace lining and the charge was used, and it is possible that some of the zirconium was lost by oxidation from the air. A dark stiff slag heavier than usual was formed on the heats to which zirconium was added. In spite of this it is difficult to understand why with an addition of some 69 points of zirconium

the recovery is no more than when 25 points are added. The carbon loss averages between 8 and 9 points, and the silicon loss between 16 and 17 points. The normal loss in these furnaces is around 6 points carbon and 5 points silicon. The loss of silicon is far less than that of the zirconium and points to the fact that zirconium is a much more powerful deoxidizer than silicon. It would also appear that zirconium is alloyed with some difficulty.

All the ingots gave normal "piping" with no indications of purging. Heats 239, 249, and C-207 showed signs of wildness in the test spoon and required additional silicon and zirconium to "kill" them. The ingots were stripped hot, buried in cinders and allowed to go cold before reheating for forging. The castings were also cooled in ashes before heat treatment.

The composition for these heats was based on the present standard composition used at the Watertown Arsenal for making centrifugal castings, namely,

	Per Cent
Carbon	0.35 to 0.45
Manganese	0.50 to 0.80
Silicon	0.15 to 0.35
Molybdenum	0.25 to 0.35
Vanadium	0.05 to 0.10

Table II
Addition and Recovery of Alloys

	CARBON		SILICON		ZIRCONIUM		VANADIUM		MOLYBDENUM	
No.	Added	Recovered	Total added as Fe-Si and in Fe-Si-Zr Alloy	Recovered	Added	Recovered	Added	Recovered	Added	Recovered
173	0.38	0.365	0.40	0.31	0	0	0.12	0.16	0.30	0.25
239	0.37	0.30	0.45	0.24	0.25	0.017	0.15	0.12	0.30	0.30
249	0.37	0.29	0.40	0.225	0.25	0.0238	0.15	0.105	0.30	0.33
263	0.43	0.335	0.36*	0.29	0.25*	0.0496	0.15	0.14	0.30	0.30
C-207	0.43	0.35	0.41	0.24	0.25	0.0204	0.15	0.10	0.30	0.27
C-209	0.43	0.34	0.41*	0.23	0.28*	0.0410	0.15	0.14	0.30	0.32
C-218	0.47	0.36	0.40	0.21	0.69	0.0265	0.20	0.16	0.30	0.29

*On these heats ferro-silicon-zirconium (Carbon 0, Manganese 1.4, Silicon 54.28, Zirconium 37.78) was used. On other heats ferrozirconium (Carbon 0.50, Manganese 0.40, Silicon 15.51, Aluminum 8.88, Titanium 4.68, Zirconium 65.2) and ferrosilicon (46.6 per cent Silicon) were used.

NOTE:—On C-218, 28 points Zirconium added as ferrosilicon-zirconium and balance as ferrozirconium. All silicon added as ferrosilicon-zirconium.

except that the vanadium was raised to 0.15 per cent, the manganese additions eliminated and zirconium added instead (except Heat 173).

Forging of Ingots

The heats to which zirconium was added showed no indications of red-shortness. Forging in one case was carried as far as 24 to 1 reduction. Normal heating and soaking times were applied to the ingots in an oil-fired furnace, the maximum temperature used being 2150 degrees Fahr. (1175 degrees Cent.) and the minimum slightly above the critical temperature. Forgeability over this range, using press and hammer, appeared to be as good as any ordinary alloy steel. Hot working was carried out under a variety of conditions and no trouble was experienced. The small ingots were worked under a hammer, the large ingot under a press. Ingot No. 173, without zirconium, was somewhat red-short and did not forge satisfactorily. Normal discards for ordnance steel of 30 per cent from top and 5 per cent from bottom were made, and disks for macro-etching top and bottom of ingots made from inside ends of discards.

Ingot With No Zirconium

The forgeability of No. 173 was unsatisfactory, it being somewhat red-short. It was forged to a 3 to 1 reduction to a 4-inch round, and a section about 8 inches long taken from the best part of the bar was used for heat treatment. The critical points for this ingot fell at 1390 degrees Fahr. (755 degrees Cent.) on heating and at 1265 degrees Fahr. (685 degrees Cent.) on cooling. The 4-inch round was normalized from 1740 degrees Fahr. (950 degrees Cent.) and then quenched in water from 1605 degrees Fahr. (875 degrees Cent.). Brinell hardness readings taken at this time showed the material to be soft at one end and hard at the other. It was next heated to 1650 degrees Fahr. (900 degrees Cent.) and quenched in water. It then Brinelled as follows, the two sets of readings being 180 degrees apart:

No. 1	side	477	418	241	255	364
No. 2	side	375	460	235	387	269

A draw at 1200 degrees Fahr. (650 degrees Cent.) gave on No. 1 side:

269 364 241 255 228

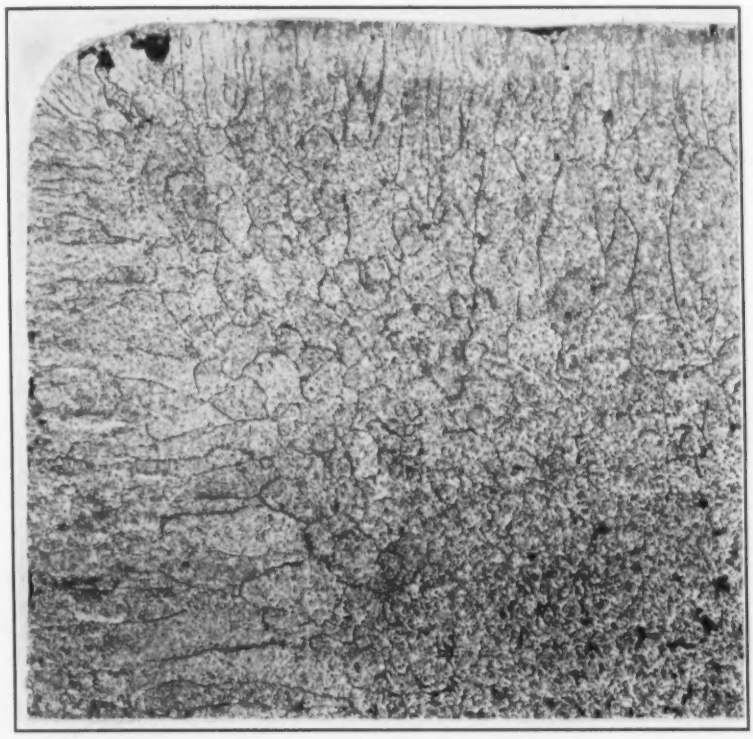

Fig. 1—Macrostructure of Top of Ingot 173 As-cast. Size 6½ Inches Square.

Quenching from 1695 degrees Fahr. (925 degrees Cent.) did not improve uniformity. After quenching from 1740 degrees Fahr. (950 degrees Cent.) the hardness was found to be:

No. 1	side	228	228	217	228	228
No. 2	side	228	241	228	241	255

This high quench gave the nearest approach to uniformity. Longitudinal tensile bars taken from the stock in the last quenched condition, without any draw, gave:

Location	Proportional Limit Lbs. per Square Inch	Tensile Strength Lbs. per Square Inch	Elongation in 2 Inches	Reduction of Area	Brinell on Bar
Center of 4-Inch Round	64,000	111,000	19.0	40.3	230
Midway between center and outside	75,000	111,500	22.0	57.2	225

The response to hardening was certainly abnormal and unsatisfactory, but the tensile properties were very good. Fig. 1 shows a

quarter section of the macro-etched disk from the top of the ingot. In respect to external appearances this ingot was as good as any made. Pronounced ingot structure is in evidence. The hot acid etch[3] discloses a selective action apparently around the grain boundaries.

Ingots Containing Zirconium

Ingots 239 and 249 were also forged 3 to 1 reduction to 4-inch rounds. Figs. 2 and 3 show respectively quarter sections of the macro-etched structure of top and bottom of No. 239, as cast, and Figs. 4 and 5 show quarter sections of the same forged 3 to 1. No. 249 was practically a duplicate of 239, and hence is not shown here. These two ingots showed a striking similarity to 173 on the etch test. A material quite rapidly attacked by the etching solution has been removed, disclosing a coarse columnar structure.

Table III shows average test results on the 4-inch rounds after heat treatment, which consisted of a normalize from 1700 degrees Fahr. (925 degrees Cent.), a water quench from 1605 degrees Fahr. (875 degrees Cent.) and a draw, followed by cooling in the furnace. The Charpy values were obtained from a tensile impact bar. To find the approximate equivalent that would be given by the small square notched Charpy bar, it is necessary to multiply these values by 0.60.

It will be noted that for the higher draw the impact value is slightly lower, while the reduction of area remains approximately the

Table III

Physical Properties of 4-Inch Round Bars From Heats 239 and 249

Longitudinal Bars	Proportional Limit Pounds per Square Inch	Tensile Strength Pounds per Square Inch	Elongation	Reduction	Charpy
Drawing Temperature 1200 Degrees Fahr.					
No. 239	77,500	101,100	23.7	61.0	31.7
No. 249	75,500	99,500	24.4	64.2	31.0
Transverse Bars					
No. 239	81,000	106,000	25.0	63.1	35.4
No. 249	82,000	107,000	24.2	62.7	30.0
Drawing Temperature 1290 Degrees Fahr.					
Transverse Bars only					
No. 239	62,000	90,000	27.1	61.5	30.5
No. 249	60,000	86,500	28.5	64.7	30.4

[3]The standard acid etch used at the Watertown Arsenal on all ordinary alloy steels is a solution of 38 per cent hydrochloric acid, 12 per cent sulphuric acid and 50 per cent water by volume. This is maintained at the simmering point for 40 minutes during immersion of the specimen.

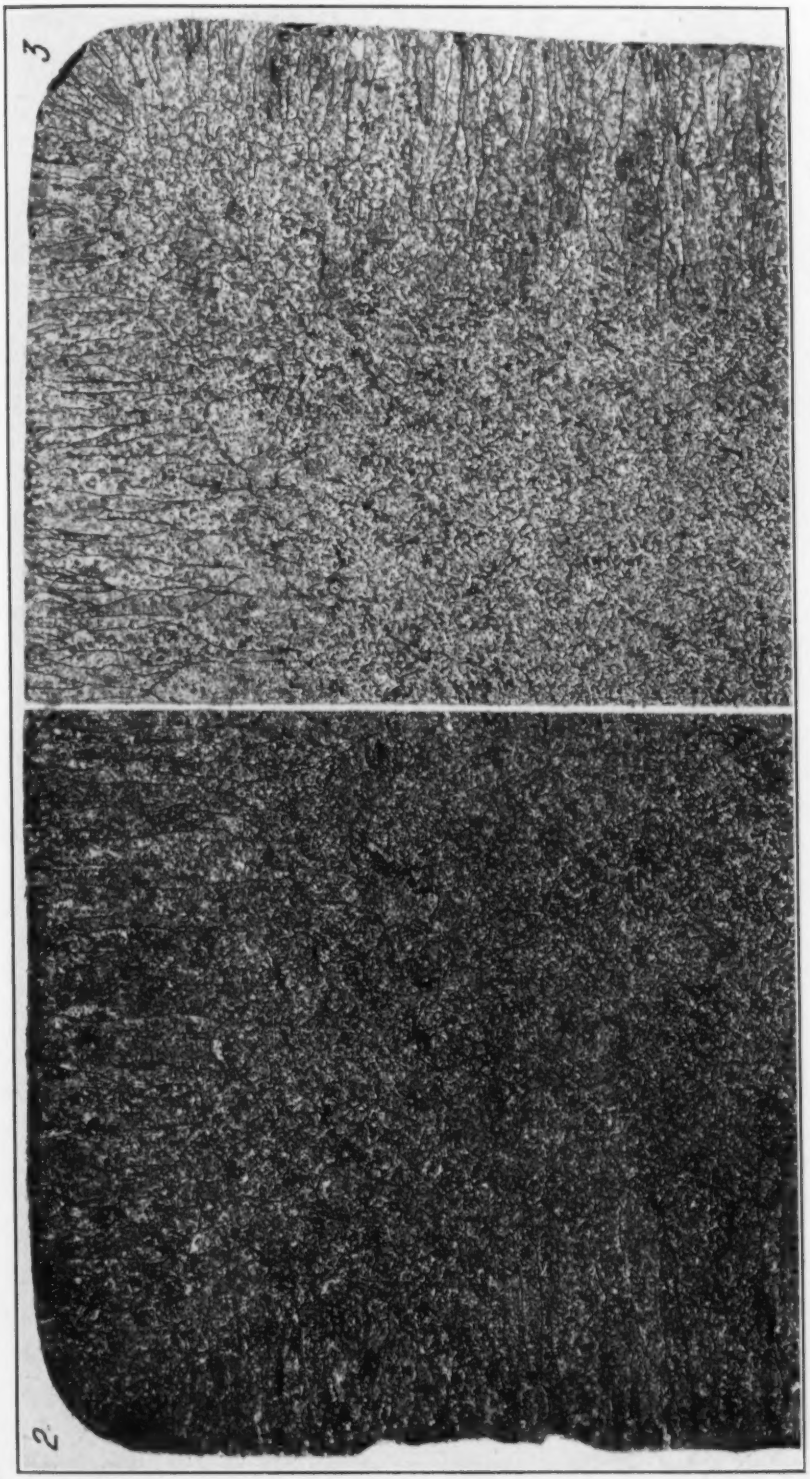

Fig. 3—Macrostructure of Bottom of Ingot 239 As-cast.

Fig. 2—Macrostructure of Top of Ingot 239 As-cast. Size 6½ Inches Square.

same and the elongation increases. Furthermore, the transverse bars appear to give as good ductility as the longitudinal bars.

Fig. 6 discloses a microstructure on test bars of 239, drawn at 1200 degrees Fahr., of fine uniform sorbite with some ferrite uniformly distributed in very small dots or areas. No segregation is

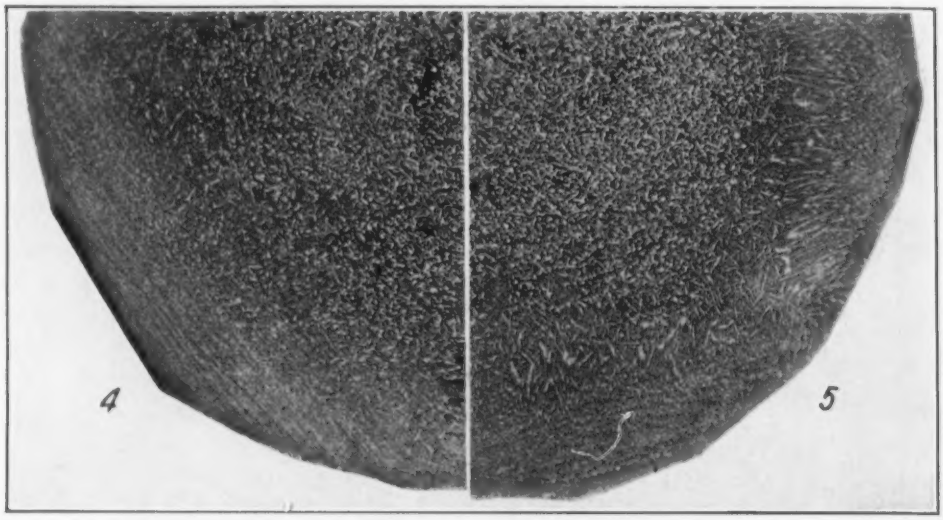

Fig. 4—Macrostructure of Forging, Heat 239 After Forging to 3 to 1 Reduction. Top of Ingot Section.
Fig. 5—Macrostructure of Forging, Heat 239 After Forging to 3 to 1 Reduction. Bottom of Ingot Section.

apparent. The specimens examined were taken from mid-wall and toward the outside of a 4-inch round after heat treatment. The structure was practically the same from center to outside.

Ingot No. 263

Inasmuch as ingots 239 and 249 gave satisfactory results, it was decided to make a 2½-ton ingot. The 2¼-ton high frequency induction furnace was used for this purpose. The ingot was cast in a 21-inch fluted mold by slightly building up the bottom. Usual discards were made and a slab taken from top and bottom and macro-etched, quarter sections of which are shown in Figs. 7 and 8. The ingot was then forged 2 to 1 reduction to a 14-inch round under a press and again slabbed at extreme ends of the forging for the etch test. Quarter sections of the macro-etched disks are shown in Figs. 9 and 10.

Some slag inclusions are in evidence at the side exterior of the ingot represented by Fig. 7. The columnar structure, so evident in

the small ingots, is absent in this case. Perhaps the much slower cooling has either caused diffusion or prevented the formation during solidification of the material so readily etched away from the grain boundaries. The slag referred to above is still in evidence in Fig. 9 and has caused a few surface imperfections. No chipping was necessary in forging the ingots containing zirconium. Fig. 9 also

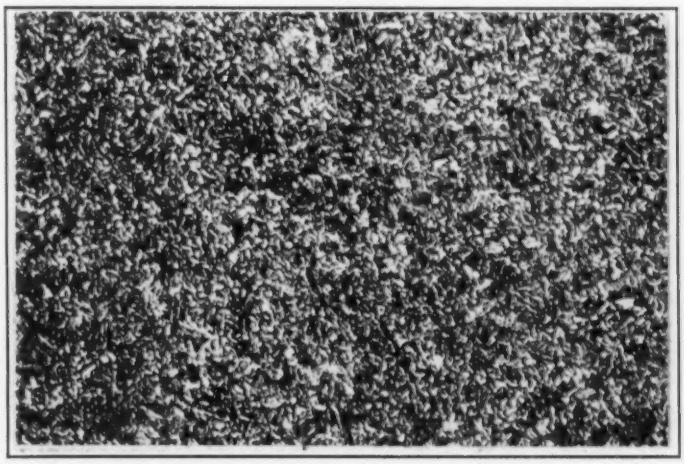

Fig. 6—Macrostructure of Forging Heat 239 Quenched at 1605 Degrees Fahr. and Drawn at 1200 Degrees Fahr. × 100.

discloses what might be taken to be a slight porosity at the center. Such was not the case. This condition was caused by the fact that the disk was cut so near the end as to leave less than ⅛-inch thickness at the center, and the acid ate completely through. Disks 4 inches long were cut from the top and bottom end of the forging and heat treated as follows for physical test:

Degrees Fahr.	Degrees Cent.	
1740	(950)	4 hours, air-cooled
1605	(875)	4 hours, water-quenched
1245	(675)	4 hours, furnace-cooled
1275	(690)	4 hours, furnace-cooled

Brinell hardnesses on the surfaces of the two disks after treatment are given in Table IV.

Transverse test bars taken at different radii from the center of the bottom disk after the draw at 1275 degrees Fahr. are shown in Table V.

The properties shown in Table V were not satisfactory. The impact values were erratic and in most cases exceedingly low. It

Fig. 8—Macrostructure of Bottom of 21-Inch Ingot (5000 pounds) Heat No. 263. As-cast.

Fig. 7—Macrostructure of Top of 21-Inch Ingot (5000 pounds) Heat No. 263. As-cast.

Table IV
Brinell Hardness Readings on Disks from Forging 263 After Treatment

Disk	Location—Center to Outside Diameter	Values After Draw at 1245 Degrees Fahr. (675 Degrees Cent.)
263—Top of Ingot	To 12 o'clock	217-241-228-235-241-241-241-241
263—Top of Ingot	To 3 o'clock	217-228-228-238-241-241-241-241
263—Top of Ingot	To 6 o'clock	217-235-241-238-241-241-241-241
263—Top of Ingot	To 9 o'clock	217-228-241-228-228-241-235-241
263—Bottom of Ingot	To 12 o'clock	228-228-231-231-228-235-235-231
263—Bottom of Ingot	To 3 o'clock	228-228-228-231-231-235-238-235
263—Bottom of Ingot	To 6 o'clock	228-228-228-235-235-235-235-228
263—Bottom of Ingot	To 9 o'clock	228-228-228-231-228-228-226-228
After Re-Draw At 1275 Degrees Fahr. (690 Degrees Cent.)		
263—Top	To 12 o'clock	192-199-196-196-199-199-196-196
263—Top	To 3 o'clock	192-196-192-190-190-194-192-196
263—Top	To 6 o'clock	192-196-196-196-199-202-202-207
263—Top	To 9 o'clock	192-202-199-196-196-196-202-207
263—Bottom	To 12 o'clock	207-196-202-196-194-196-196-196
263—Bottom	To 3 o'clock	207-199-196-199-194-196-196-187
263—Bottom	To 6 o'clock	207-202-202-202-196-202-194-212
263—Bottom	To 9 o'clock	207-196-194-187-196-196-202-207

Table V
Test Results on Forging 263—2 to 1 Reduction

No. of Bar	Location	Proportional Limit Pounds per Square Inch	Tensile Strength Pounds per Square Inch	Per Cent Elongation	Per Cent Reduction of Area
1	1.5-inch radius	51,000	83,500	21.5	46.2
1	1.5-inch radius	51,000	84,000	22.5	51.9
2	3.5-inch radius	51,000	83,500	23.0	57.2
2	3.5-inch radius	53,000	88,500	23.5	57.2
3	6.0-inch radius	63,000	93,000	24.0	59.8
3	6.0-inch radius	54,000	87,000	23.5	62.3

CHARPY

No.	Location	Value Foot-Pounds
BC	Exactly in center of disk	9.2 6.5
BC 1	1-inch radius	3.7 5.0
BC 2	2-inch radius	5.9
B 1	2.5-inch radius	8.2 7.9
B 2	5.0-inch radius	23.6 5.8
B 3	6¾-inch radius	26.1 23.9

appeared necessary either to develop a better heat treatment, or to give more hot working, or both in order to obtain higher impact values. After a study of the microstructure it was felt that mass-effect was a serious factor and that a more drastic quench would be necessary for the comparatively low carbon used in these ingots.

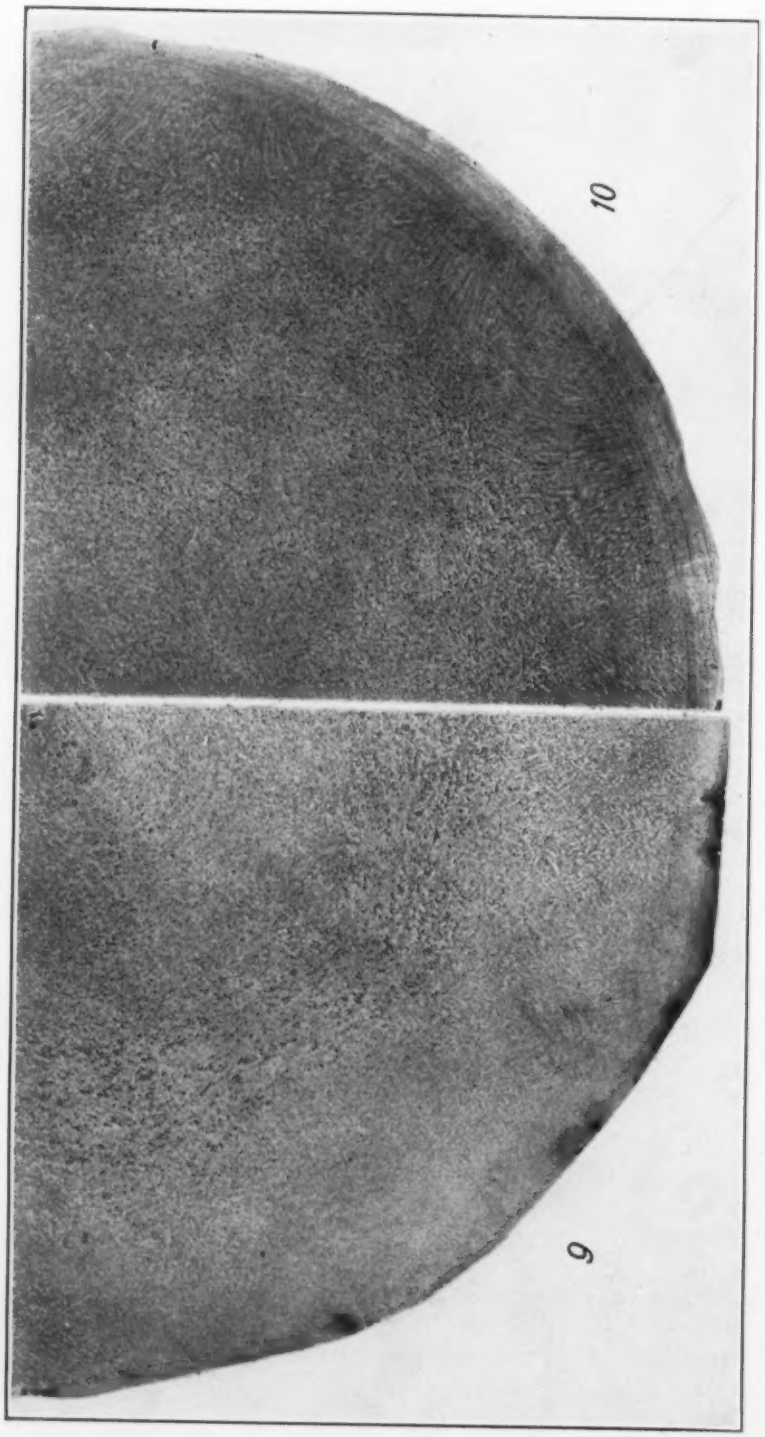

Fig. 9—Macrostructure of Top of Ingot No. 263 After Forging 2 to 1 Reduction. (14-Inch Round).
Fig. 10—Macrostructure of Bottom of Ingot No. 263 After Forging to 2 to 1 Reduction. (14-Inch Round).

Fig. 11 shows the microstructure of a test bar, taken on the 5-inch radius after the draw at 1275 degrees Fahr., to be fairly fine uniform sorbite with a considerable amount of ferrite in fairly uniformly distributed areas. There are also traces of ingotism. The structure is not that of a well hardened steel but approaches more nearly to a normalized condition. It is to be noted that this structure was found on the 5-inch radius of a 14-inch disk 4 inches thick.

A 1¼-inch thick slab approximately 14 inches long by 5½ inches wide was cut from the end of the bottom disk of 263 used above, and this slab cut transversely through the center into two pieces each approximately 7 inches x 5½ inches x 1¼ inches. One of these pieces was quenched in water from 1830 degrees Fahr. (1000 degrees Cent.) and the other from 2010 degrees Fahr. (1100 degrees Cent.). Brinell readings taken over the surfaces approximately 1 inch apart showed for the lower quench values ranging from 241 to 512, with an average of 430, and for the higher quench 321 to 512, with an average of 462. Greater uniformity was given by the higher quench. Both pieces were drawn at 1245 degrees Fahr. (675 degrees Cent.) for two hours and cooled in the air. Brinell impressions taken as before the draw gave values for the piece quenched at 1830 degrees Fahr. from 241 to 269, with an average of 263, and for the slab quenched at 2010 degrees Fahr. 241 to 286, with an average of 264. Both pieces were again drawn at 1275 degrees Fahr. (690 degrees Cent.) and cooled in the air. Hardnesses taken as before gave for the lower quench values from 217 to 241, with an average of 234, and for the higher quench values from 223 to 248, with an average of 233. It will be seen that the resulting uniformity and hardness is about the same for the two quenches. The results of transverse test bars taken from these pieces after the last draw are given in Table VI. The radii indicated show the distance of the test bars from the center of the original 14-inch forging.

The microstructure of specimens taken on the 3.5-inch radius, indicated in Table VI, quenched from 1830 degrees Fahr., is shown in Fig. 12. It is rather coarse-grained, strongly cleavaged sorbite. No appreciable amount of ferrite is found in the grain boundaries. Some segregation in small areas is apparent. The quenching has been drastic but it has not greatly improved the impact strength. The structure of the specimens quenched from 2010 degrees Fahr. was practically the same as that shown in Fig. 12, except slightly coarser.

Table VI
Additional Tests on 263, 2 to 1 Reduction

Treatment	Radii of Specimen Inches	Proportional Limit Pounds per	Tensile Strength Square Inch	Elongation in 2 Inches	Reduction in Area	Charpy Foot-Pounds
	1.5	77,000	102,500	20.0	59.8	...
	2.5	.....	.....	...	...	10.8
Quenched at 1830 degrees Fahr.	*3.5	79,000	106,500	20.0	62.3	...
(1000 degrees Cent.)	5.0	.....	.....	...	...	23.7
Drawn at 1275 degrees Fahr.	6.0	87,000	110,000	20.0	59.8	...
(690 degrees Cent.)	6.75	.....	.....	...	...	9.6
	1.5	80,000	104,000	20.5	59.8	...
	2.5	.....	.....	...	...	7.1
Quenched at 2010 degrees Fahr.	*3.5	79,000	105,000	19.5	57.2	...
(1100 degrees Cent.)	5.0	.....	.....	...	...	8.1
Drawn at 1275 degrees Fahr.	6.0	83,000	110,000	20.0	62.3	...
(690 degrees Cent.)	6.75	.....	.....	...	...	11.8

*For micro examination.

Tensile test bars taken from 263 forged 2 to 1 reduction were satisfactory, but the Charpy was low. To ascertain if further hot working would raise the impact value a section of 263, 14 inches in diameter, was reforged in steps, 3 to 1, 4 to 1 and 5 to 1 reduction. Only the results of the 2 to 1 and 5 to 1 tests are given in this paper.

A 4-inch disk (14 inches in diameter), 2 to 1 reduction, and a 4-inch disk (9.5 inches in diameter), 5 to 1 reduction, were treated as follows—using an oil-fired furnace:

1920 degrees Fahr. (1050 degrees Cent.)	4 hours, quenched in water to black heat (below critical), then cooled in air.
1740 degrees Fahr. (950 degrees Cent.)	4 hours, quenched in iced brine.
1200 degrees Fahr. (650 degrees Cent.)	4 hours, cooled in air.

Brinell hardness, taken on four quadrants from center to outside at the surface of the 14-inch disk after treatment showed an

Table VII
Tests on 263—5 to 1 Reduction

	Direction	Radii Inch	Proportional Limit Lbs. per Sq. In.	Tensile Strength Lbs. per Sq. In.	Elongation	Reduction of Area	Charpy	Brinell Hardness on Bar
5 to 1 Reduction	Longitudinal	1.5	82,000	114,000	21.0	54.8	6.1	228
		2.5	84,000	116,000	21.0	55.8	13.3	228
		3.5	86,000	116,000	21.0	58.4	17.8	235
	Transverse	1.5	82,000	116,000	16.0	34.8	8.3	228
		2.5	82,000	116,000	18.0	43.4	6.5	228
		3.5	82,000	116,000	20.0	51.5	7.0	228

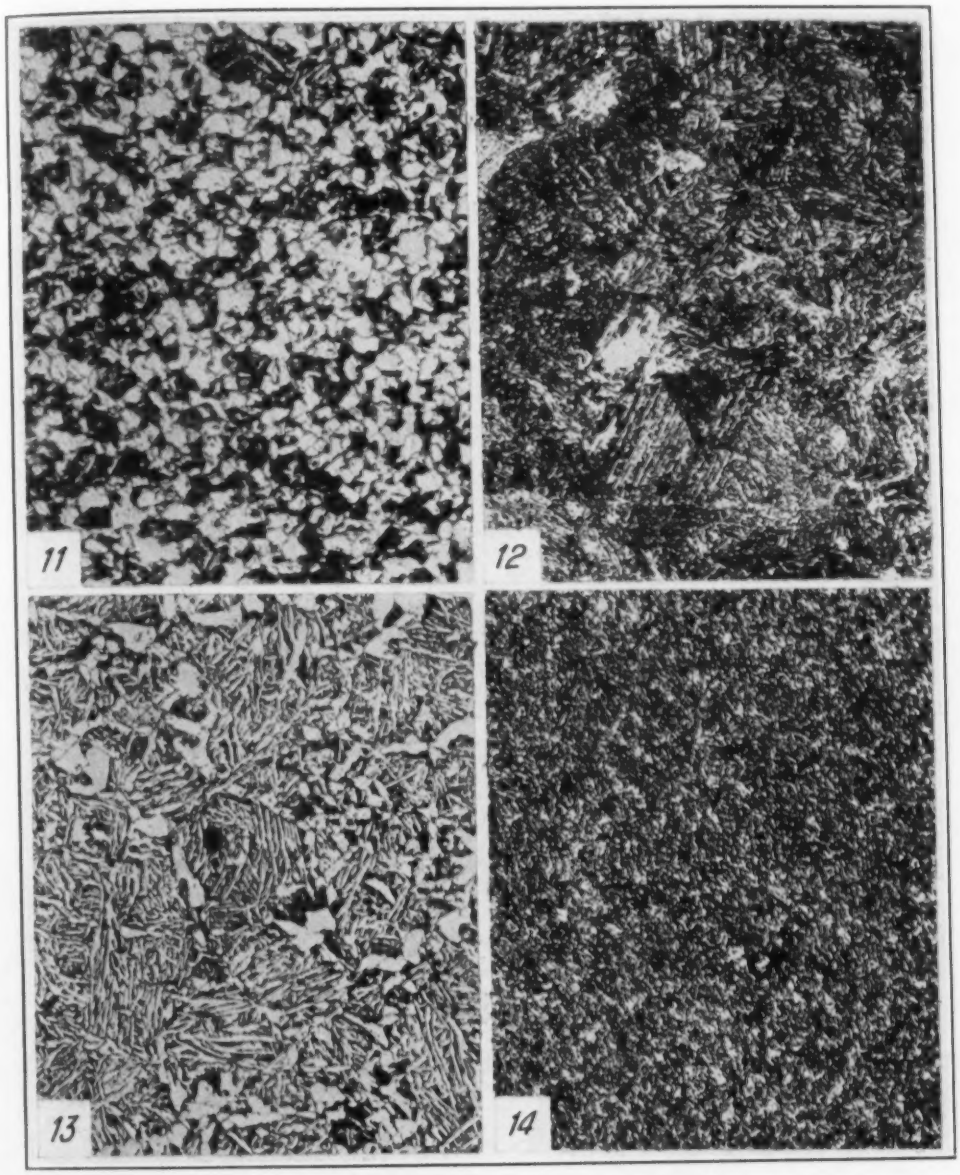

Fig. 11—Microstructure of Forging No. 263, 2 to 1 Reduction. Quenched at 1605 Degrees Fahr., Drawn at 1275 Degrees Fahr. 4-Inch Section. × 100.
Fig. 12—Microstructure of Forging No. 263, 2 to 1 Reduction. Quenched at 1830 Degrees Fahr., Drawn at 1275 Degrees Fahr. × 100.
Fig. 13—Microstructure of Forging No. 263, 5 to 1 Reduction. Quenched at 1740 Degrees Fahr., Drawn at 1200 Degrees Fahr. × 100.
Fig. 14—Microstructure of Forging No. 263, 15 to 1 Reduction. Quenched at 1605 Degrees Fahr., Drawn at 1275 Degrees Fahr. × 100.

average of 257 over a range of 248 to 262. The 9.5-inch disk gave an average hardness of 251 over a range of 241 to 269.

Longitudinal and transverse test bars taken from the center of the 9.5-inch slab on radii indicated are shown in Table VII.

A test bar on the 2.5-inch radius of the 9.5-inch section (the Charpy bar giving 6.5 value) had the structure shown in Fig. 13. Again it is coarse strongly cleavaged sorbite, slightly banded with a considerable amount of ferrite in the grain boundaries.

A section of the 4-inch disk that was forged 5 to 1 and heat treated was redrawn at 1275 degrees Fahr. (690 degrees Cent.) for two hours and cooled in water. It then showed a uniform hardness of 241 Brinell on the surface. Transverse test bars taken from the center of the section on radii indicated are shown in Table VIII.

Table VIII
Tests on 263—Drawn at 1275 Degrees Fahr. (5 to 1 Reduction)

Radii Inch	Proportional Limit Lbs. per Sq. In.	Tensile Strength Lbs. per Sq. In.	Elongation	Reduction of Area	Charpy	Brinell on Bar
1.5	78,000	108,000	22.0	50.8	11.8	223
2.5	80,000	110,000	23.0	54.1	23.0	217
3.5	84,000	112,000	22.0	58.4	21.3	217

A block from the 9.5-inch disk was forged to a ¾-inch diameter and a 12-inch length, a reduction equivalent to 15 to 1. This bar and a ¾-inch thick radial section from the 9.5-inch disk were treated 1 hour together as follows, using an electric muffle furnace:

1740 degrees Fahr. (950 degrees Cent.), air cool
1605 degrees Fahr. (875 degrees Cent.), water quench
1200 degrees Fahr. (650 degrees Cent.), air cool

Both bar and radial section showed uniform hardness after the quench of 512 Brinell and after the draw, 302 Brinell. They were redrawn at 1275 degrees Fahr. (690 degrees Cent.) for 1 hour and cooled in air. The bar then had a uniform hardness of 248 and the radial section 241. Longitudinal test bars gave values indicated in Table IX. The properties given are satisfactory.

Table IX
Tests on 263—15 to 1 and 5 to 1 Reduction

Specimen	Proportional Limit Lbs. per Sq. In.	Tensile Strength Lbs. per Sq. In.	Elongation	Reduction of Area	Charpy	Brinell
¾ Inch Bar	98,000	116,000	24.0	66.6	44.2	228
15 to 1	96,000	112,000	25.0	64.8	47.0	217
Reduction					44.4	228
¾ Inch Slab	—1.5 Inch	Radius	....	....	33.7	217
5 to 1	—2.5 Inch	Radius	....	....	34.2	217
Reduction	—3.5 Inch	Radius	....	....	35.6	228

The microstructures of test specimens from the ¾-inch bar and from the ¾-inch slab (specimen on the 2.5 inch radius) were the best of any examined in the whole series of those treated. They were quite alike and showed very fine-grained uniform sorbite with no segregations or free ferrite. The typical structure obtained from the ¾-inch bar (15 to 1) is shown in Fig. 14. The slab that received the 5 to 1 reduction had perhaps a slightly coarser structure than that shown in Fig. 14, and this is revealed in the lower impact value. The excellent results from the treatment given in this case as compared with the specimen represented by Fig. 12 indicate that the quenching temperatures used for the latter were too high, thus leaving unduly large grains.

A 4-inch disk from forging 263 (9.5-inches round) forged 5 to 1 reduction, was heated to 1920 degrees Fahr. (1050 degrees Cent.) for 4 hours and cooled in water to black, and then cooled in air. It was again heated to 1740 degrees Fahr. (950 degrees Cent.) for four hours and cooled drastically in water to a black. A quadrant taken from the disk was heated at 1605 degrees Fahr. (875 degrees Cent.) for 2 hours and quenched in cold water. It then showed a hardness on the surface of 387 Brinell. After drawing at 1320 degrees Fahr. (715 degrees Cent.) for 2½ hours and cooling in water it gave a hardness of 228 Brinell. Test bars taken from the quadrant after this treatment gave results as follows from transverse specimens taken in the center of the 4-inch section on radii indicated:

Radii Inch	Proportional Limit Lbs. per Sq. In.	Tensile Strength Lbs. per Sq. In.	Elongation	Reduction of Area	Charpy	Brinell on Bar
1.5	72,000	96,000	27.0	64.8	43.4	196
2.5	70,000	96,000	26.0	64.8	37.9	187

These properties are satisfactory for gun forgings.

Cold Work and Aging Tests

In order to get some indication as to how steel with low manganese would respond to cold work and aging treatment, a section of 239 was forged to a 1⅜-inch round bar. It was then normalized from 1740 degrees Fahr. (950 degrees Cent.), water-cooled from 1605 degrees Fahr. (875 degrees Cent.), and drawn at 1200 degrees Fahr. (650 degrees Cent.), followed by furnace cooling. The treatment was accomplished from an oil-fired furnace. Time at heat in each

operation was 2 hours. The bar as treated was approximately 36 inches long. After the quench it showed a fairly uniform hardness from end to end of 444 Brinell, and after the draw a hardness of 241 to 255 Brinell. Test bars taken from the two ends after the draw gave the following results:

	Proportional Limit Pounds per	Tensile Strength Square Inch	Per Cent Elongation	Per Cent Reduction of Area	Charpy
A-End	84,000	108,000	25.0	63.8	45.8
A-End	86,000	110,000	25.0	65.1	41.5
B-End	92,000	116,000	23.0	65.5	40.4
B-End	90,000	116,000	24.0	62.8	47.5

The bar was then machined, and stretched in a testing machine to give 6 per cent cold work.

The bar was then cut in two and the B-end low-temperature annealed for 2 hours at 570 degrees Fahr. (300 degrees Cent.) followed by cooling in the furnace. The yield strength after cold work was 106,000 pounds per square inch on the center bar and 105,000

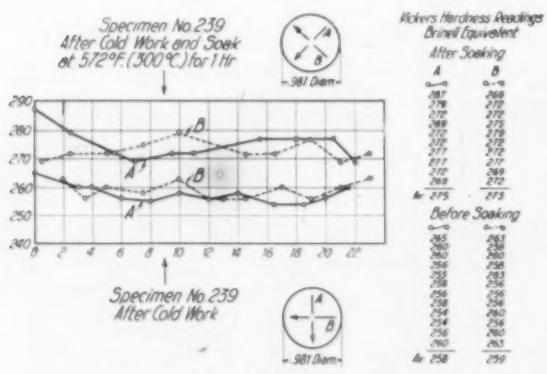

Fig. 14a

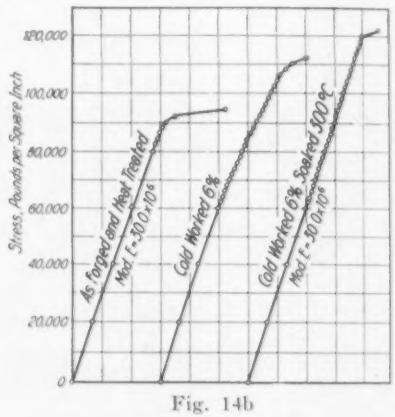

Fig. 14b

Fig. 14a—Curves Showing Hardness Values from Low Temperature Treatment on Disk.
Fig. 14b—Stress-Strain Curves.

pounds per square inch on the other two bars. The curves in Fig. 14a indicate the increase in hardness from the low temperature treatment on disk.

It is well known that in general cold-worked steel does not show a right line stress-strain curve. Soaking at from 480 degrees Fahr. (250 degrees Cent.) to 755 degrees Fahr. (400 degrees Cent.) usually gives a right line stress-strain curve and a definite proportional limit. Fig. 14b shows the stress-strain curve (a) as heat treated, (b) as cold-worked and (c) as cold-worked and aged.

The microsctructure of a specimen taken from the bar after

heat treatment showed a fine uniform sorbite with some ferrite in evenly distributed very small dots. Slight banding was noted in some of the longitudinal views.

The microstructure in this case was satisfactory and the physical properties shown in the preceding table are regarded as excellent. These specimens, however, had a great deal of hot work and were quenched in a small section, hence should give good results. It would appear from the values obtained that the limit in cold working for the particular heat treatment given had been reached as the proportional limit after aging is the same as the tensile strength. This material appears to respond to cold working and aging in about the same degree as steels with normal manganese.

Centrifugal Casting of Steels Without Manganese

The composition of the steel used for making cannon by the centrifugal process at the Watertown Arsenal contains only one strategic material, namely, manganese. One prime reason for carrying on this work was to ascertain if a steel could be made without manganese, cast centrifugally, and heat treated to meet the gun forging requirements.

The casting for the 75 M/M Pack Howitzer was selected. As cast it weighs about 690 pounds and has a breech diameter of 8 inches, a muzzle diameter of 4.5 inches, and is 81 inches long. The bore as cast is approximately 1 inch.

Casting C-207 was the first made without using manganese. Fig. 15 is a macro-etched quarter disk taken 6 inches from the muzzle end. The pronounced similarity of the muzzle structure on the outer annuli to that of ingot 239 is apparent. At the muzzle will be seen two different concentric structures, an inner and an outer. This structure is peculiar to this composition. For example, note Fig. 16 is a quarter disk from the muzzle end of a pack howitzer made with practically identical composition to that of C-207, except that it contained 0.70 per cent manganese and no zirconium. Casting conditions such as speed of rotation, rate of pour, and temperature of the chill mold were also the same. The casting shown with normal manganese gave satisfactory transverse physical properties, indicated in Fig. 17, breech end.

Sections 7 inches long were cut from breech and muzzle end of C-207 for heat treatment, which consisted of:

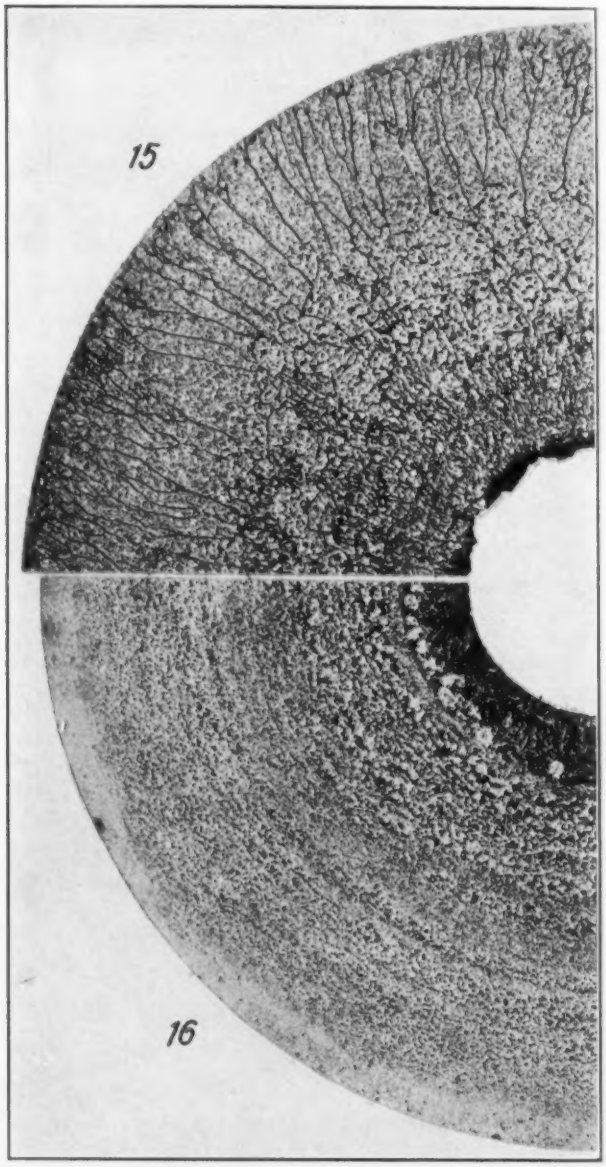

Fig. 15—Macrostructure of Centrifugal Casting C-207, 6 Inches from Muzzle End, As-cast.
Fig. 16—Macrostructure of Centrifugal Casting, 6 Inches from Muzzle End, 0.70 per cent manganese, No Zirconium. (See Fig. 17).

1740 degrees Fahr. (950 degrees Cent.)—3 hours, air chill
1607 degrees Fahr. (875 degrees Cent.)—4 hours, water quench
1290 degrees Fahr. (700 degrees Cent.)—4 hours, furnace cool

Transverse test bars taken on radii indicated across the wall at breech and muzzle gave very low ductility as shown in Table X.

The fractures of most of the test bars giving poor ductility especially at the muzzle end, showed large conchoidal silvery sur-

Table X
Tests on C-207

Specimen		Radii Inch	Proportional Limit Lbs. per Sq. In.	Tensile Strength Lbs. per Sq. In.	Elongation	Reduction of Area	Charpy
Breech	No. 1	1	66,000	100,000	9.0	20.6	24.0
	No. 1	1	68,000	96,000	6.0	20.6	17.9
							22.0
	No. 2	1.604	62,000	94,000	16.0	30.8	....
	No. 2	1.604	54,000	78,000	3.0	20.6	....
	No. 3	2.25	58,000	88,000	29.0	55.8	23.6
	No. 3	2.25	56,000	90,000	27.0	54.8	27.9
							26.5
	No. 4	3.15	52,000	90,000	23.0	46.2	29.2
	No. 4	3.15	64,000	88,000	22.0	34.0	19.4
							19.0
Muzzle	No. 1	1	68,000	92,000	9.0	17.0	4.2
	No. 1	1	68,000	80,000	9.0	17.0	6.1
							10.1
				Broke at			
	No. 2	1.604	66,000	66,000	4.0	0	2.2
	No. 2	1.604	68,000	66,000	0	0	2.6
							1.2

faces, apparently the outlines of the grains shown in the macro-etch. Disks cut from these same heat treated sections showed identical structure to that of the untreated casting. Evidently the treatment had failed entirely to refine the coarse structure. It was necessary to resort to high normalizing temperatures to break down and diffuse the material that etched away so rapidly from the grain boundaries. This material under the microscope gave the appearance of oxide streaks. With a drastic normalize it seemed to disperse throughout the grain. A number of temperatures were tried, and while possibly not the best, an air chill from 2100 degrees Fahr. (1150 degrees Cent.) followed by one at 2010 degrees Fahr. (1100 degrees Cent.) seemed to break down all traces of the original coarse grain structure. For example Figs. 18 and 19 show quarter sections of disks taken 2 inches in from the ends of casting C-209 (covered below) as cast and Fig. 20—a half section—the muzzle disk after the following treatment:

2100 degrees Fahr. (1150 degrees Cent.)—2 hours, air cool
2012 degrees Fahr. (1100 degrees Cent.)—2 hours, air cool
1607 degrees Fahr. (875 degrees Cent.)—2 hours, water quench
1275 degrees Fahr. (690 degrees Cent.)—3 hours, furnace cool

This treatment applied to the disk in Fig. 15 gave:

Specimen	Radii	Proportional Limit Pounds per	Tensile Strength Square Inch	Elon-gation	Reduction of Area
Muzzle No. 1	1 inch	82,000	106,000	21.0	30.8
Muzzle No. 2	1.604 inches	86,000	108,000	22.0	49.2

It was noted that the coarse structure seemed to be more pronounced at the quickly chilled muzzle end. Also the test coupons taken from the molten bath showed similar structure around the edges where the metal was chilled very quickly. It seemed as if by solidifying the metal more slowly a better structure might be ob-

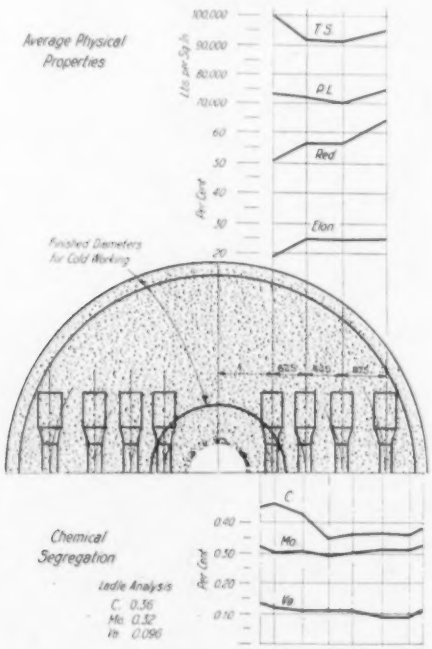

Fig. 17—Macrostructure, Composition, and Physical Properties of Breech End, Centrifugal Casting, Normal Manganese.

tained. The pouring time for C-207 was 24 seconds. Another casting was poured at a slower rate to see if the structure would be improved thereby. Casting C-209 was made by pouring the metal into the mold in 77 seconds. A speed of 1100 revolutions per minute, the same as used for C-207, was maintained. Slightly more zirconium was also added to this heat.

Disks from the ends (see Figs. 18 and 19) showed practically identical structure to C-207. The change in rate of pour did not seem to affect the macrostructure at all. Sections 7 inches long were again taken from breech and muzzle ends and center of casting and treated as follows:

2100 degrees Fahr. (1150 degrees Cent.)—4 hours, air cool
2012 degrees Fahr. (1100 degrees Cent.)—4 hours, air cool
1607 degrees Fahr. (875 degrees Cent.)—4 hours, water quench
1290 degrees Fahr. (700 degrees Cent.)—5 hours, furnace cool

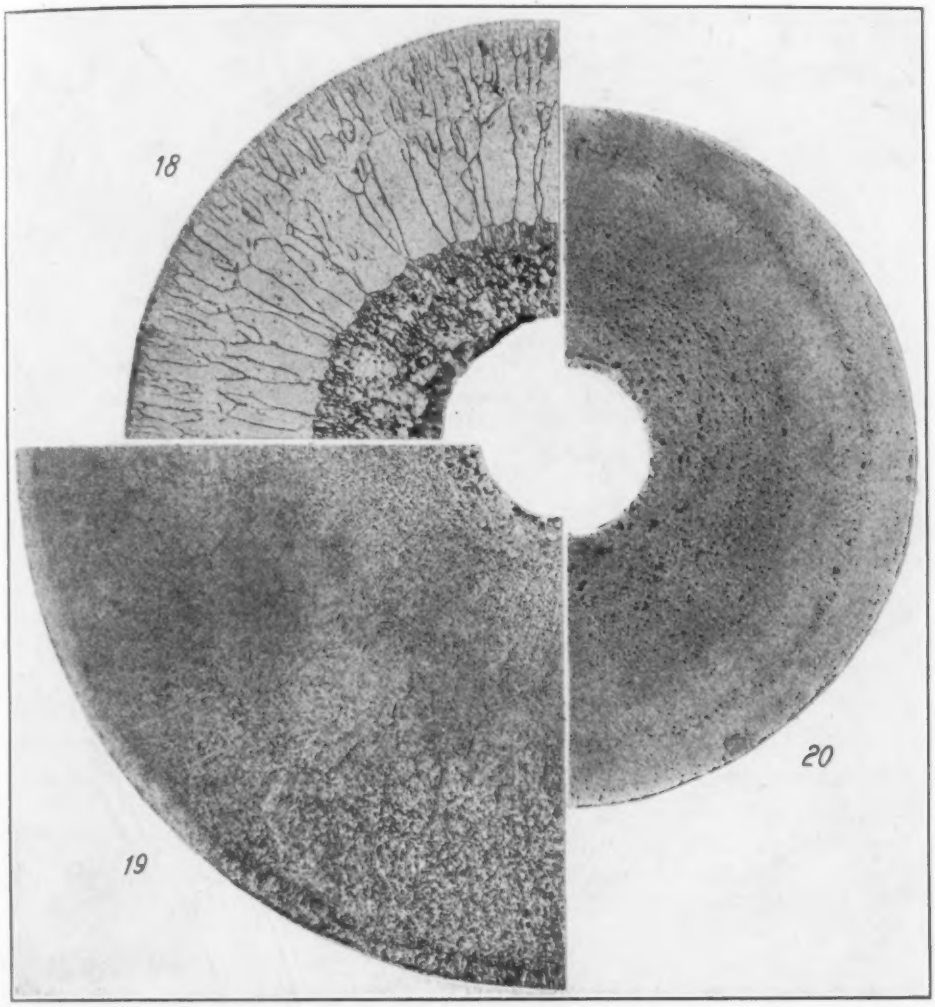

Fig. 18—Macrostructure of Centrifugal Casting C-209, 2 Inches from Muzzle End, As-cast.

Fig. 19—Macrostructure of Centrifugal Casting C-209, 2 Inches from Breech End, As-cast.

Fig. 20—Macrostructure of Disk in Fig. 18 After Following Treatment: 2100 Degrees Fahr. 2 Hours Air-cool; 2010 Degrees Fahr. 2 Hours Air-cool; 1605 Degrees Fahr. 2 Hours Water-quenched; 1275 Degrees Fahr. 3 Hours, Furnace-cool.

This treatment gave on transverse bars taken on different radii from the center of the sections the results shown in Table XI:

The strength given by the tensile bars is satisfactory but the ductility is erratic and low. A few of the fractures still exhibited traces of the silvery spots. Possibly the soaking time at heat was not quite long enough to completely diffuse the coarse structure.

Another casting, C-218, was made, using a heat to which a total of 69 points of zirconium was added. Macro-etched disks 2 inches

Table XI
Additional Physical Strength Tests on C-207

Specimen		Radii Inch	Proportional Limit Lbs. per Sq. In.	Tensile Strength Lbs. per Sq. In.	Elongation	Reduction of Area	Charpy
Breech	No. 1	1	58,000	96,000	14.0	17.7	19.7
	No. 1	1	70,000	98,000	11.0	14.4	17.0
							15.9
	No. 2	1.604	64,000	94,000	17.0	38.6	....
	No. 2	1.604	62,000	92,000	16.0	21.2	....
	No. 3	2.25	64,000	90,000	24.0	61.4	20.4
	No. 3	2.25	62,000	88,000	25.0	49.2	20.1
							20.5
	No. 4	3.15	64,000	94,000	23.0	54.1	35.8
	No. 4	3.15	66,000	94,000	23.0	52.0	35.2
							32.3
Muzzle	No. 1	1	70,000	104,000	15.0	26.8	25.2
	No. 1	1	78,000	108,000	17.0	39.8	19.7
							17.8
	No. 2	1.604	78,000	108,000	25.0	58.4	6.1
	No. 2	1.604	74,000	100,000	14.0	30.2	5.1
							12.3
Center	No. 1	1	64,000	92,000	15.0	35.4	4.8
	No. 1	1	64,000	92,000	19.0	29.4	19.4
	No. 2	1.604	66,000	90,000	14.0	27.4	20.2
	No. 2	1.604	64,000	90,000	17.0	34.0	14.5
	No. 3	2.25	68,000	94,000	22.0	51.1	25.3
	No. 3	2.25	68,000	94,000	23.0	50.4	27.1

in from breech and muzzle ends, shown in Figs. 21 and 22, disclose an excellent macrostructure at the breech but again the typical coarse grain at the muzzle. Two sections, each 6 inches long, were taken from breech and muzzle ends and treated as follows:

2100 degrees Fahr. (1150 degrees Cent.)—3 hours, air cool
2012 degrees Fahr. (1100 degrees Cent.)—3 hours, air cool

One section from each end was quenched in water from 2010 degrees Fahr. (1100 degrees Cent.) and the other section from each end from 1605 degrees Fahr. (875 degrees Cent.). Disks taken from each of these four sections gave no macro-trace of the coarse grain structure. Fig. 22 shows the structure at the muzzle after treatment. The condition represented in Fig. 23 has disappeared.

After a draw at 1290 degrees Fahr. (700 degrees Cent.) for four hours, followed by furnace cool, test bars taken on different radii from the center of the sections gave the results shown in Table XII.

Brinell hardnesses taken on the heat treated sections from breech and muzzle ends from which test bars were cut gave results shown in Table XIII.

Specimens for micro-examination were taken from the heat treated sections of C-218 ("L" sections), on the 2-inch radius at the

Table XII
Tests on C-218

Specimen No.*	Radius Inch	Proportional Limit Lbs. per Sq. In.	Tensile Strength Lbs. per Sq. In.	Elongation	Reduction of Area	Charpy	Brinell on Bar
HB1	1	86,000	114,000	12.0	20.5	22.3	...
HB2	1	Broken in Mach.		....	....	17.9	...
						22.0	
HB3	2	84,000	110,000	21.0	52.6	26.5	...
						26.2	
HB4	2	78,000	114,000	19.0	50.4	29.8	223
HB5	3	88,000	110,000	22.0	61.4	21.2	...
						24.7	
HB6	3	78,000	114,000	22.0	57.4	28.0	217
HM1	1	90,000	118,000	16.0	37.2	11.1	...
						12.0	
HM2	1	92,000	120,000	13.0	25.4	10.6	...
HM3	1.5	88,000	116,000	19.0	43.4	19.1	...
						22.2	
HM4	1.5	92,000	122,000	18.0	39.8	23.4	241
LB1*	1	58,000	94,000	20.0	27.4	23.6	...
LB2	1	68,000	94,000	15.0	26.0	27.1	...
						15.5	
LB3	2	66,000	90,000	25.0	48.6	30.1	...
						31.8	
LB4	2	60,000	90,000	25.0	56.8	30.4	179
						34.0	
LB5	3	66,000	90,000	27.0	63.2	40.5	...
LB6	3	68,000	92,000	25.0	55.2	37.1	179
LM1	1	74,000	96,000	11.0	18.4	17.2	...
						15.2	
LM2	1	74,000	96,000	13.0	28.1	13.0	...
LM3	1.5	76,000	102,000	24.0	50.4	28.6	...
						32.9	
LM4	1.5	76,000	100,000	23.0	56.8	28.3	187

*Specimens marked "H" quenched from 2010 degrees Fahr. Specimens marked "L" quenched from 1605 degrees Fahr.

Table XIII
Brinell Readings on C-218 Sections

	Breech				Muzzle			
Diameter of Circle	After 1100° C Quench	After Draw	After 875° C Quench	After Draw	After 1100° C Quench	After Draw	After 875° C Quench	After Draw
1⅝	303	238	264	194	...	...	...	...
2⅜	295	241	265	195	365	264	257	209
3⅛	286	236	254	189	374	263	257	208
3⅞	284	235	245	187	364	253	259	211
4⅝	269	234	242	187	...	...	...	...
5⅜	281	234	241	189	...	...	...	...
6⅛	277	234	241	193	...	...	...	...
6⅞	268	239	243	191	...	...	...	...

breech and on the 1.5-inch radius at the muzzle. Figs. 24 and 25 (muzzle and breech) indicate the structure to be fine uniform sorbite with some ferrite evenly distributed in small dots. There appears to be no segregation. The breech, as was to be expected, is slightly coarser than the muzzle. It is interesting to compare Figs. 6 and 24. The former represents a forging (3 to 1 reduction) and the latter a

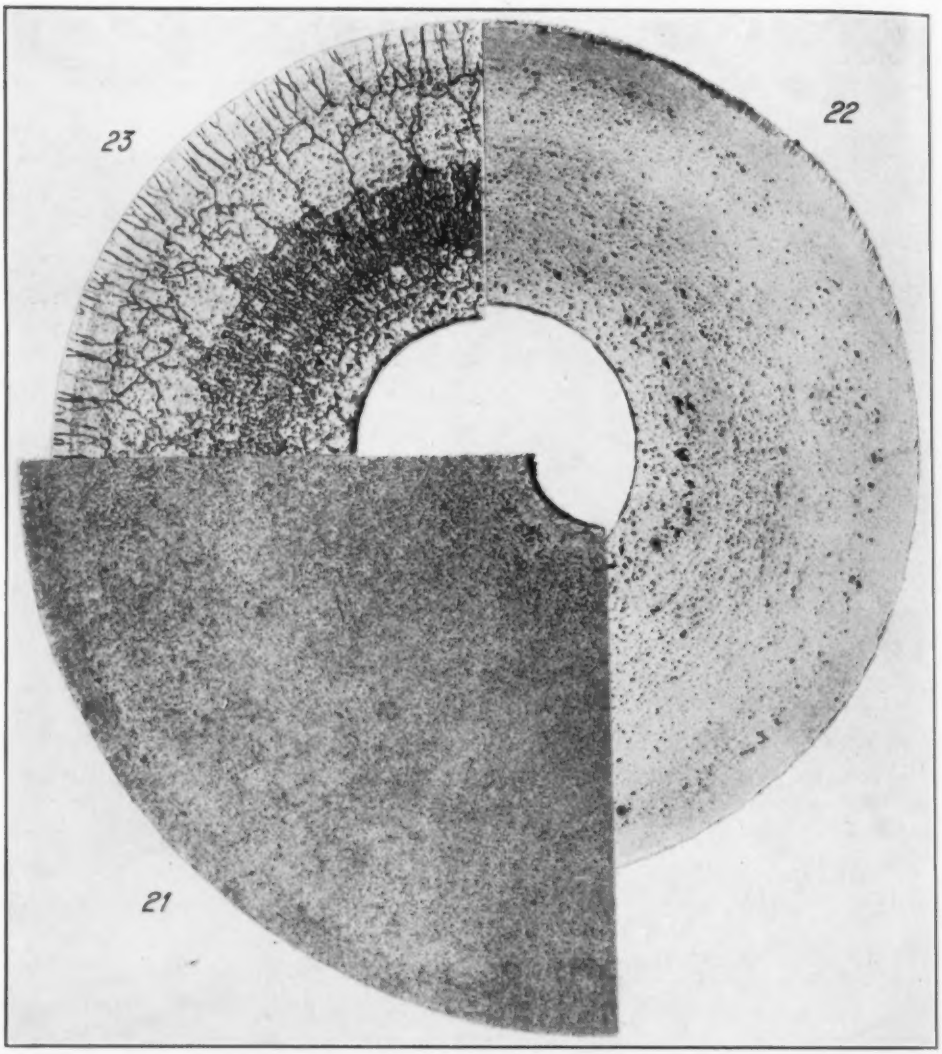

Fig. 21—Macrostructure of Centrifugal Casting C-218, 2 Inches from Breech End, As-cast.
Fig. 22—Macrostructure of Centrifugal Casting C-218, 2 Inches from Muzzle End, As-cast.
Fig. 23—Macrostructure of Centrifugal Casting C-218, Muzzle End, After Heat Treatment.

centrifugal casting. The casting appears to have a somewhat better structure than the forging.

It is pointed out that test bars taken on the 1-inch radius of C-218 are in metal that would be removed from a casting before it goes into service. Disregarding these inner bars, which should be done in this case, the other test specimens for the lower quench meet the gun forging requirements, and are entirely satisfactory.

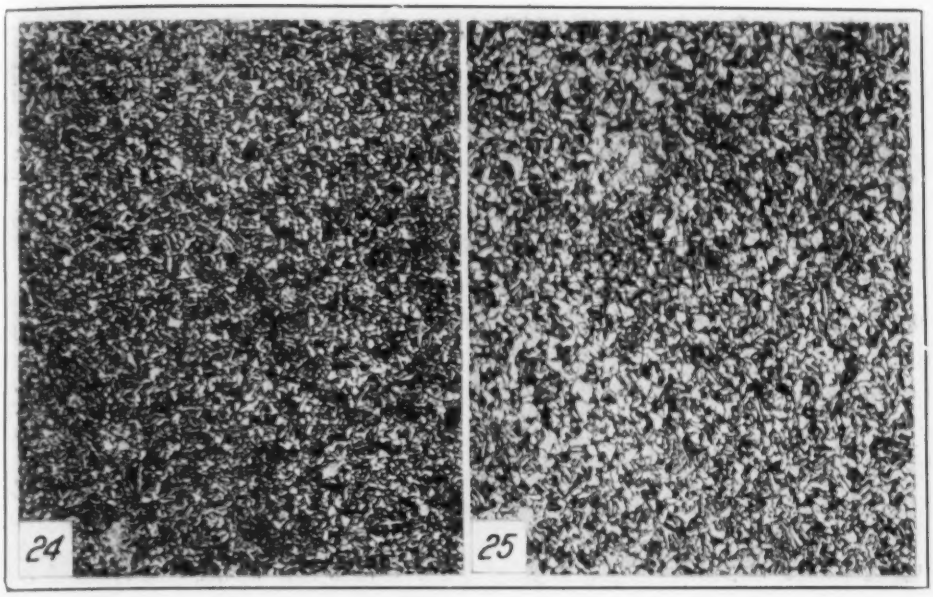

Fig. 24—Microstructure of Centrifugal Casting, C-218, Muzzle End, After Heat Treatment (Lower quench) × 100.
Fig. 25—Microstructure of Centrifugal Casting C-218, Breech End, After Heat Treatment (Lower quench). × 100.

Inclusion Analyses

In order to ascertain the relative cleanliness of the castings and forgings containing low manganese as compared with good commercial stock, a dirt count somewhat along the methods followed by Kinzel and Crafts[4] is in progress. Preliminary results indicate that these steels are slightly higher in dirt content than was high grade open-hearth bar stock. The inclusions being counted are 0.005 millimeters or larger in diameter. The distribution follows in general the curve shown in Fig. 5 article by Kinzel and Crafts. A great many small inclusions below 0.005 millimeters are apparent. This work is being done by Miss Mary Norton under the supervision of Dr. H. H. Lester in the Arsenal Laboratory.

Tests for Abnormality

Grossmann pointed out in his Campbell Memorial Lecture presented in 1930 ("Oxygen in Steel") that abnormality probably is caused by the absorption of oxygen during the carburizing treatment, and he gives evidence to show that the degree of abnormality

[4]A. B. Kinzel and Walter Crafts, "Inclusions and Their Effect on Impact Strength of Steel," Technical Publication No. 402, American Institute of Mining and Metallurgical Engineers, 1931.

is proportional to the amount of oxygen absorbed. Samples of 239, 249 and 263 were carburized and examinations made in accordance with the McQuaid-Ehn test for abnormality. The specimens examined indicated abnormality, but not to any degree expected from the steel with such low manganese content. The case obtained was about 0.071-inch deep, maximum carbon estimated to be 0.90 per cent. There was no appreciable hypereutectoid case.

In this connection it is to be noted that the great number of hardness readings recorded, all of which are not reported herein, showed reasonable uniformity (except in the case of 173) to be expected from stock heat treated in ordinary commercial production furnaces with no undue precaution taken to insure uniformity. Further work is in progress to ascertain if the stock reported on will show abnormality when carburized in the absence of oxygen.

General Discussion

The small ingots all showed decided columnar structure, more pronounced in those containing zirconium. Centrifugal castings also showed this condition at the muzzle ends, where the metal was chilled more rapidly than at the breech ends where the structure was in general satisfactory as judged by the macro-etch test. However, even at the breech it required higher normalizing treatments than ordinarily given to standard centrifugal castings to develop satisfactory physical properties. In the 5000-pound ingot this columnar structure was not apparent. What the material is that etches away so rapidly from around the grains or dendrites in the castings and small ingots has not been definitely determined. Under the microscope it appears as oxide streaks or "chains." The fact that this condition is found to a certain extent in 173 seems to indicate that it cannot be attributed altogether to zirconium or any of its compounds. It has been suggested that possibly it was a nitride. It is necessary either to hot work the steel or resort to extra high normalizing treatments to completely diffuse the material from around the grain boundaries. It appears that the higher the heat the more thorough the diffusion; it was found, however, that 2100 degrees Fahr. (1150 degrees Cent.) was about the maximum safe temperature to use on centrifugal castings. 2190 degrees Fahr. (1200 degrees Cent.) showed conditions bordering on incipient fusion. The fact that the castings at the muzzle ends and the small ingots all give selective etching apparently

around the grains, and the castings at the large ends and the large ingot all fail to show this action points to the conclusion that the precipitation of this material is greatly dependent upon the rate of solidification and that slow cooling after this point tends to cause its diffusion. It appears to be soluble in gamma iron, as shown by the response to the diffusion treatment, which in the larger sections is accomplished to a great extent during and after solidification. It is known that steel castings (normal manganese) will show higher ductility and more uniform structure if permitted to remain in the sand molds until practically cold. This permits a certain amount of diffusion of the material in the local areas last to solidify, rich in carbon and impurities. Cooling had proceeded too fast and too far to permit this to take place on the muzzle ends of the castings and in the small ingots when they were stripped from the molds.

The total oxygen content of the low manganese steels as determined by the vacuum fusion method is within reasonable limits. The oxygen found averaged well with steels containing normal manganese. Based on the amount present no difficulty should be expected from oxygen. However, it would be necessary to know how and in what form the oxygen reported occurs before suspicions in that direction can be eliminated. Possibly another element, reported in larger than average amounts, is a worse offender than oxygen in the zirconium steels. Zirconium has quite an affinity for nitrogen and tends to form nitrides, some of which pass into the slag and others remain in the steel as yellow crystals or flakes. It is possible that the presence of nitrogen is partly responsible for some of the low impact values reported in this paper.

Inclusion counts and residue analyses disclose nothing so far to indicate that the low manganese steels are materially dirtier than the average of good stock. However, in some specimens examined under the microscope there was evidence of dirt or nonmetallics in a very finely dispersed state. No. 263 appeared dirtier in this respect than the other steels and this may partly explain also the lower impact values obtained on this ingot. Kinzel and Crafts[5] pointed out that "factors other than visible inclusions have an important influence on the dynamic strength of steels, and are comprised in the term 'body'," and that "body is markedly influenced by minute inclusions whose

[5]A. B. Kinzel and Walter Crafts, "Inclusions and Their Effect on Impact Strength of Steel," Technical Publication No. 402, American Institute of Mining and Metallurgical Engineers, 1931.

size is less than 0.005 millimeter." While factors comprised in the term "body" are probably the cause for the low impact values realized in some of the tests on 263, the strength, longitudinal and transverse, disclosed by the tensile test bars was satisfactory and the ductility indicated by elongation and reduction was excellent. Usually it is possible to predict in a fair degree the Charpy values to be realized once the ductility on the tensile bar is known. The tensile bars of 263 would indicate a high impact, but in fact the values were remarkably low.

Centrifugally cast guns containing normal manganese show structures and physical properties equivalent to forgings of equal composition and size. The low-manganese castings maintain this distinction in a remarkable degree, even at the inferior-appearing muzzle end. C-218 shows a structure in Figs. 24 and 25 at the muzzle and breech very similar to forging 239 in Fig. 6. It appears to be as fine and uniform as the forging. The structure at the muzzle end of C-207 compares very favorably with 239. In strength and ductility the castings, when properly treated, have responded as well as the forgings.

The physical properties, except the low Charpy results on 263, compare favorably with those of similar alloy steels containing normal manganese. Hardness after the quench ran from about 387 on the 4-inch thick disks to 600 Brinell on ¼-inch test bars. Penetration of hardening throughout the heavier pieces was about the same as would be expected from ordinary alloy steel with normal manganese and equivalent carbon content. During the tests a number of drastic water quenches were made and no tendency to checking or cracking was noted. In practically every case after proper treatment had been applied the tensile test bars exhibited the fibrous, cup-and-cone type of fracture characteristic of high quality steel. No "crow-feet" or other surface imperfections indicative of dirty steel could be found at or near the fractures. Likewise the impact specimens, except from 263, gave usual shear areas at the breaking point. The fractures on the low charpies from 263, taken adjacent to tensile bars that gave high elongation and reduction, were coarse crystalline with no shear in evidence. Careful examination under the binocular microscope disclosed a network of crystals of variable size and lead to the belief that the low impacts were possibly due to a very fine oxide or nitride precipitate along the crystal faces or around the grain boundaries.

It appears that low-manganese steels are much lower in resistance to ordinary rusting and corrosion than similar steels with normal manganese. Oxidation under heat is faster also, and when the large (14-inch) disks were being polished after treatment for hardness readings it was necessary to remove more metal than usual in such cases to obtain representative hardness.

Conclusions

1. Based on the rather limited test data given in this paper, it appears that steel can be made in high frequency induction furnaces without using manganese, and heat treated to meet the gun forging requirements for strength, either in forgings or centrifugal castings.

2. Only two prime cleansing agents were used, namely silicon and zirconium. While zirconium is apparently necessary to prevent red-shortness, it is possible that other deoxidizers could be used to good advantage. The elements carbon, silicon, vanadium, and zirconium could all be increased, it is believed, with beneficial results.

3. High normalizing treatments are essential. They should be followed by drastic quenching, especially on heavy sections, from ordinary hardening temperatures. Such treatment seems to give maximum diffusion at least expense in grain growth.

4. Low manganese steels respond to cold working and aging in about the same degree as steels with normal manganese.

5. The results reported herein illustrate the value of the Charpy test for steels to be subjected to severe dynamic stresses in service.

DISCUSSION

Written Discussion: By Albert Sauveur, Gordon McKay Professor of metallurgy and metallography, Harvard University, Cambridge, Mass.

Captain Ritchie's excellent description of the manufacture of steel made without manganese is of much interest. The fear that if we had no manganese we could not manufacture steel should have been dispelled many years ago when the American Rolling Mill Company started the manufacture of ingot iron without manganese. If steel so low in carbon that it is generally referred to as ingot iron can be produced without manganese and without the use of zirconium, it is not readily understood why steel higher in carbon and manganese-free requires the use of zirconium. Several heats of steel substantially free from manganese and containing from 0.10 to 1.00 per cent carbon and over were made by the American Rolling Mill Company and the difficulty in forg-

ing reported by the author for his ingot 173 was not encountered, nor were these steels found to be abnormal when subjected to hardening. I do not wish to be understood by these remarks, however, to claim that zirconium is not beneficial in making steel without manganese.

One should be impressed by the results reported by the author showing the physical properties of transverse test bars to be fully equal to the properties of longitudinal bars. While this may be ascribed partly to the low oxygen content of the zirconium treated steel, I believe it to be due in a large measure to the very low phosphorus content present.

The Watertown Arsenal should be warmly congratulated for its important contributions in recent years to the metallurgical art, namely the judicious use of cold work deformation and of heat treatment to increase the elastic limit of guns, the manufacture of centrifugally cast steel guns, the application of X-rays to the detection of flaws in castings and of defects in welds and now the manufacture of steel without manganese.

Written Discussion: By C. H. Herty, Jr., director of research, metallurgical advisory committee, Carnegie Institute of Technology, Pittsburgh.

Captain Ritchie's paper is very interesting in that it again shows that zirconium will replace manganese in preventing the red-shortness of steels during forging or rolling. This point was previously brought out by Feild and others some time ago. It would seem that a research on the use of zirconium in connection with steels containing from 0.20 to 0.40 per cent manganese would have been of much more practical benefit, because almost all scrap contains the above amount of manganese as a minimum and almost all steels made by the open-hearth process will contain this amount of manganese without any ferro-manganese addition. From a technical standpoint, this work is therefore of interest. From a more practical standpoint it would appear to have little significance.

One question which arises in this connection is that of the accessibility of zirconium in comparison with quite large tonnages of low-grade manganese ores. A method of using the low-grade manganese ores in this country has been demonstrated in U. S. Bureau of Mines Report of Investigations 3107 entitled "A Practical Method of Solving the Emergency Manganese Problem."

Finally, there is one statement which should not go unchallenged: The last sentence in the paper reads "No surface imperfections indicative of dirty steel could be found." This is certainly misleading. It has been my experience that surface imperfections are traced more to the lack of complete deoxidation and to improper heating of the ingot than to any other factor. I have seen large amounts of very dirty steel which gave a beautiful surface on rolling.

Written Discussion: By A. B. Kinzel, Union Carbide and Carbon Research Laboratories, Long Island City, N. Y.

Captain Ritchie is deserving of our highest commendation on the very vigorous and satisfactory way in which he has attacked the problem of making steels without manganese. As Captain Ritchie explains, this is peculiarly a war problem, and has long been recognized as such. That the solution of this problem could be effected by the use of zirconium in place of manganese was first appreciated by Dr. F. M. Becket, who published a paper on "Some Effects

of Zirconium in Steels" in the *Transactions* of the American Electrochemical Society in 1923. Since then further work on the general effect of zirconium in steel has been carried out and described by A. L. Feild of the Union Carbide and Carbon Research Laboratories, Inc., of which Dr. Becket is president, and more recently Dr. Becket has described a whole series of manganese-free zirconium-treated steels in both plain carbon and alloy variety, in his address before the meeting of the Iron and Steel Division of the A. I. M. E. in February, 1931. This exposition of the problem, together with a description of his experimental results, has been published in *Mining and Metallurgy* for May, 1931.

It has been our privilege to cooperate with Captain Ritchie of the Watertown Arsenal since the inception of the work described in this paper. Part of this cooperation consisted of running check tests on the materials produced by Captain Ritchie, as well as supplying manganese-free steel which we had made in the course of the same type of work.

The ingots which we made were necessarily smaller than those made at Watertown, and possibly for this reason improved physical properties were obtained. For example, the heat containing 0.22 per cent carbon, 0.004 per cent manganese, 0.35 per cent silicon, 0.10 per cent phosphorus, 0.33 per cent molybdenum, 0.12 per cent vanadium and 0.19 per cent zirconium, oil-quenched from 900 degrees Cent. and drawn at 600 degrees Cent. in 1-inch rounds and tested as 0.505 inch specimens, resulted in 81,000 pounds per square inch yield point, 96,000 pounds per square inch ultimate strength, 26 per cent elongation in 2 inches, 65 per cent reduction of area, 52 foot pounds Izod, and a Brinell hardness of 179. Again, a chromium-nickel steel containing 0.44 per cent carbon, 0.008 per cent manganese, 0.36 per cent sulphur, 0.006 per cent phosphorus, 1.03 per cent chromium, 3.50 per cent nickel and 0.15 per cent zirconium, heat treated as above mentioned, showed 120,000 pounds per square inch yield point, 135,000 pounds per square inch ultimate strength, 21 per cent elongation in 2 inches, 60 per cent reduction of area, 46 foot-pounds Izod, and 269 Brinell. A chromium steel containing 0.37 per cent carbon, 0.028 per cent manganese, 0.39 per cent silicon, 0.026 per cent sulphur, 0.013 per cent phosphorus, 0.92 per cent chromium, and 0.32 per cent zirconium, gave values in the heat treated condition, of 75,000 pounds per square inch yield point, 107,000 pounds per square inch ultimate strength, 24 per cent elongation in 2 inches, 63 per cent reduction of area, 21 foot-pounds Izod, and a Brinell of 212. A chromium-vanadium steel containing 0.43 per cent carbon, 0.018 per cent manganese, 0.31 per cent silicon, 0.032 per cent sulphur, 0.022 per cent phosphorus, 0.94 per cent chromium, 0.26 per cent vanadium and 0.13 per cent zirconium, in the heat treated condition gave values of 117,000 pounds per square inch yield point, 139,000 pounds per square inch ultimate strength, 18 per cent elongation in 2 inches, 52 per cent reduction of area, 22 foot-pounds Izod and 300 Brinell. From these figures it is apparent that the use of zirconium in the absence of manganese results in most excellent properties not only in steels which may be particularly selected for ordnance use, but also for automotive and engineering uses in general. It is specially noteworthy that no difficulty whatever was experienced in the hot working of these materials even when the sulphur was reasonably high.

In connection with the physical properties above noted, it should be particularly remarked that the impact resistance is improved by the zirconium addition. Other work at these laboratories indicates that this effect is even more important than might be supposed from a study of the figures themselves. Impact in engineering work is an indication not so much of the energy required to break a given specimen, as of the tendency to failure in service by means of notch formation and notch propagation. We have some indications to the effect that the high impact values obtained in the zirconium-treated steels are due primarily to the effect of zirconium on the reduction of the rate of notch formation and of propagation of cracks. Thus, zirconium steels may well find engineering uses even though manganese is readily available.

Captain Ritchie's results with centrifugal castings are particularly interesting, and his method of eliminating undesirable cast structure is well conceived and readily applicable in connection with this problem. In fact, the entire work comprises a most happy combination of theory and practice, and we trust that we may have more papers of this character in the not too distant future.

Oral Discussion

C. H. HERTY:[1] After going out to the Arsenal, I was very much impressed with one thing, and it seems to me that in view of the fact that this is an emergency proposition some study should certainly be carried out on why the zirconium efficiencies in this process are so low. You are paying a high compliment when you say they are "efficiencies." Almost all of the zirconium put in is lost, and it seems to me that if we are going to make zirconium replace some manganese, we certainly ought to find out how to use less zirconium than is required in some of this work on account of the very heavy zirconium losses.

H. H. LESTER:[2] Captain Ritchie has made an excellent beginning toward the solution of a vexing problem. The Ordnance Department is interested in the question, "Can steel be made without manganese or with reduced quantities of manganese?" A solution of this problem will necessarily advance our knowledge of steel making and point the way to better steels. Captain Ritchie's results indicate that by the use of zirconium it may be possible either to eliminate manganese or to reduce the percentages required. His data suggest also that possibly better steels may result from the use of zirconium. This latter deduction agrees with findings by Dr. Becket and his co-workers in the Union Carbon and Carbide Laboratories, and agrees with conclusions arrived at by Guy[3] who points out the possibility of better drawing and rolling qualities in these steels.

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[1]Director of research, metallurgical advisory committee, Carnegie Institute of Technology, Pittsburgh.

[2]Physicist, Watertown Arsenal, Watertown, Mass.

[3]"The Use of Zirconium Alloys in the Manufacture of Steels"—Ph. Guy, Aciers Speciaux, Metaux et Alliages, 1931, pages 509-521.

It is really not a new fact that steel can be made without manganese. One can use pure iron or iron very low in phosphorus, sulphur, and carbon and from this by the simple addition of carbon come out with excellent steel. This, however, is not the problem that interests the War Department. The real problem is to make good steel from commercially available raw materials, in standard furnaces, and under practical foundry conditions. The contribution toward the solution of this problem has been a real achievement for which Captain Ritchie deserves great credit.

The writer had opportunity to study many of the steels of this investigation under the microscope. The object of the study was to determine the role played by zirconium in the manganeseless steels. The details of the study can hardly be given in the brief compass of this discussion. Some conclusions arrived at may be of interest. They are given below.

We found that low carbon steels made without zirconium and without manganese were very abnormal, according to a McQuaid-Ehn test, and gave other metallographic evidence of high oxygen content. Subsequent study led to the conclusion that some of the oxygen may have entered the steel during heat treating operations.

Abnormality and evidence of oxygen content diminished with increasing carbon content and were not important for high carbon contents. These observations more or less verify results obtained by Grossmann and reported in his last year's Campbell Memorial Lecture, in which he ascribed to manganese the property of preventing the absorption of oxygen at temperatures in the heat treating range. They are consistent also with Ziegler's results presented before this present convention in which he shows lowered oxygen content with increasing carbon.

With zirconium added the steels showed a tendency toward abnormality with evidences of high oxygen content. The Bureau of Standards analysis of some of the steels examined, however, showed very low oxygen. The specimens for gas analyses did not have opportunity to absorb oxygen in heat treating operations, while the specimens examined under the microscope did have such opportunity. It is probable that the oxygen evident in the steels examined was absorbed during heat treating operations. In view of the fact that Ziegler found (in the paper referred to above) that highly refractory oxides are not completely reduced in the vacuum fusion method of gas extraction, it is possible that the Bureau's analyses did not get all of the oxygen held in the zirconium oxides that were present in these steels. However, such oxygen being fixed could not have contributed to abnormality. In view of these facts the conclusion seems warranted that zirconium does not have the property ascribed to manganese of tending to prevent oxygen absorption at forging and heat treating temperatures.

A centrifugally cast gun tube, C207, showed columnar radial grains. These were dendritic in character. On macro etching the acid ate deeply into the boundaries of these grains, penetrating in many places from ¼ to ⅓ of an inch. These boundary lines represented interdendritic segregations that could not be diffused by ordinary heat treatments. Partial diffusion was obtained by quenching from elevated temperatures, but occasional undiffused

spots could be found even after holding a ¾-inch thick slab for two hours at 1200 degrees Cent. Some of the materials in or near the dendritic boundaries were undoubtedly oxides and nitrides of zirconium. However, much of it was a material that responded to a sodium picrate etch and must have been a zirconium carbide. This carbide appearing as it does as a dendritic segregate must be insoluble in the steel even at temperatures close to the melting point. It apparently does not diffuse readily in the gamma phase. Its presence may account for the easy acid attack of the steel along the dendritic boundary lines.

In trying to diffuse the dendritic boundary segregates a new property of these steels was discovered. That is, when quenched from 1180 degrees Cent. the metal showed Brinell hardness values of from 500 to 600 BHN, at the same time the steel was tough and malleable. One piece was bent about 20 degrees without breaking or showing surface cracks.

There were numerous inclusions. Many of these were in the form of lemon yellow crystals that were taken to be zirconium nitride. The formation of the relatively large nitride crystals suggests crystallization of Zr_3N_2 from the molten bath. This in turn indicates a high melting point for zirconium nitride and a low solid solubility of the nitride. This point is an important one since it tends to show that zirconium is an excellent agent for removing nitrogen from solution in the steel.

Other inclusions appeared as finely divided dark masses of irregular form and rounded masses of a dove gray color that resembled manganese sulphide in appearance. The black masses were taken to be zirconium oxide and the rounded gray masses to be zirconium sulphide particles.

Oxide inclusions were found along the dendritic boundary lines and also at other places. Nitride crystals were also found both in and out of the dendritic boundaries.

There is evidence to support the idea that nitrides, oxides and carbides are formed in the order named. Zirconium carbides seem particularly hard to diffuse.

There was little or no evidence to indicate that any zirconium was in solution. Observed inclusions were sufficient to account for about all the zirconium found by analysis. In a private communication Dr. Jordan of the Bureau of Standards called attention to the work of Sykes[4] who found that zirconium would dissolve in pure iron up to 0.3 per cent.

The experience at the Arsenal indicates that zirconium does not enter into solution in the alloy steels such as Captain Ritchie made in anything like this proportion. The most reliable analyses indicated zirconium contents in the finished steel in the order of 0.02 per cent and this content could be largely accounted for by the presence in the steel of segregated zirconium oxides and nitrides. In view of this fact it would appear that the presence of carbon or of alloy constituents lowers the solubility of iron for zirconium.

If our observations are accurate it is evident that the chief or sole function of zirconium in Captain Ritchie's steel is that of a scavenger. It is equally effective with manganese in removing oxygen and sulphur and in addi-

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[4]"The Alloys of Zirconium," *Journal*, Institute of Metals, Vol. XXXIX, 1928, p. 173.

tion removes nitrogen. Improved rolling and drawing qualities that have been observed in zirconium treated steels may be tied up with this superior scavenging action.

Author's Closure

The discussion presented by Professor Sauveur is appreciated. Steel for ordnance requiring rigid specifications for physical properties not only should forge and roll satisfactorily but it must, among other things, give necessary transverse tensile and Charpy values after heat treatment. An attempt has been made to show that a steel of suitable quality and composition, containing little or no manganese, can be produced in quantity and forged commercially; and further, that this steel can be treated to give strength and ductility for safety in guns under normal service conditions.

The interest shown by Professor Sauveur in this work was most encouraging. Further research will undoubtedly show improvements in quality and composition of product and disclose for guns a simpler steel than that adopted for this investigation.

As stated in the paper, one of the reasons for carrying out this investigation was to ascertain if steel melted without manganese in the coreless induction furnace could be poured into centrifugal castings and heat treated to meet gun forging requirements. The centrifugal process has been established as one of the standard methods for making cannon. The induction furnace is admirably adapted to melt steel for centrifugal castings. Due to the economies resulting from the centrifugal process, and in view of present melting technique with the induction furnace it is probable that the present practice of using relatively pure melting stock will continue. However, it is believed reasonable to assume that if suitable steel can be made with little or no manganese the presence of some 0.20 to 0.40 per cent of this element would not be detrimental. The research suggested by Dr. Herty on steel containing 0.20 to 0.40 per cent manganese has been in mind since the beginning of the work reported in this paper. The present program anticipates that work will be done on such steel melted not only in induction furnaces but also in the open-hearth, with a view of determining the practicability of utilizing the low-grade manganese ores and scrap, referred to by Dr. Herty, in making ordnance steels. It is believed that the use of zirconium will be of great value in producing such steels low in manganese. So far as the author has been able to ascertain at this writing there are ample stocks of zirconium ores widely scattered throughout the United States.

Dr. Herty challenges the statement, "No surface imperfections indicative of dirty steel could be found." This statement occurs at the end of the "Extracts from a paper for A. S. S. T. Convention, 1931," which were published in the September, 1931, issue of *Metal Progress,* page 118. The sentence in the paper (under General Discussion), from which the statement in question was extracted, occurs in connection with a description of the tensile test bar fractures and reads, "No 'crow-feet' or other surface imperfections indicative of dirty steel could be found at or near the fractures." It should be clear that this statement refers to breaks or tears on the broken test pieces,

not to the surface of forged or rolled stock. It has been the experience of the author in the examination of a great many broken test bars that the fractures give in practically every case a fair indication of the quality of the steel, and that those bars that show numerous checks and tears on the stems are in most cases representative of dirty steel. It might be inferred from Dr. Herty's statement, "Surface imperfections are traced more to the lack of complete deoxidation and to improper heating of the ingot than to any other factor," that he does not regard the lack of complete deoxidation as a contributing factor in producing dirty steel. Dr. Herty and his fellow-workers deserve great credit for the work they have done toward the solution of the emergency manganese problem.

The data furnished by Dr. Kinzel are appreciated. As stated in the paper, Dr. Becket and others working under his direction in the laboratories of the Electro Metallurgical Company have rendered invaluable assistance toward the solution of the emergency manganese problem. It is hoped that they will continue their excellent work sufficiently to show that equally satisfactory results may be obtained with larger ingots.

It was the good fortune of the writer to have been associated with Dr. Lester during the investigation reported in this paper. The writer is greatly indebted to him for his assistance and valuable suggestions. His discussion, directed toward the more technical side of the subject, adds important data to the paper and points to the solution of some of the questions that were presented. It is known that Dr. Lester has much more valuable and interesting material bearing on this subject, and it is hoped he will find occasion to publish it in the near future.

In conclusion, the author again desires to express appreciation to those who took part in the discussions, and to all whose cooperation and assistance made possible the execution of the work.

TOOL STEEL FROM THE CONSUMER'S STANDPOINT

By H. G. Keshian

Abstract

This paper discusses the existing methods of selecting tool steels. The author advocates that the properties of tool steels such as wearing, fatigue, impact, non-changing, shrinking, heat resisting and other characteristics, be accurately determined and made available to the consumer in definite and numerical values, in the same manner that the tensile, heat and acid resisting, impact and fatigue properties of S.A.E. and other steels are available to the designers of machinery, buildings, automobiles and aircraft.

TO the engineer or technical man the thought of metals presents a different picture than it does to the average man who, ordinarily, thinks of metals either in their general properties or by their mere names. Iron or steel, copper or brass, aluminum or zinc, mean to him metals that are hard or soft, white or red, malleable or brittle; while the selection of metals for his various needs is generally controlled by what others have been using or what custom dictates.

An engineer or technical man, however, will generally think of metals in terms of their definite chemical or physical properties and the work of selection means to him a mental process of scrutinizing the properties under consideration. These may be physical strength, ability to conduct heat or electricity, chemical or structural stability under certain conditions, resistance to shock, wear or corrosion, adaptability to hot or cold work, etc. In other words, the name, chemical composition or structural state of a metal are only important to him insofar as they indicate certain definite quantitative properties.

Now, what does tool steel mean to the consumer? In the selection of tool steel does he evaluate the different kinds of steel in terms of their specific properties or by their mere names?

Many mechanical and structural engineers and architects have at their command valuable data pertaining to the physical and chemical properties of a variety of structural, heat and acid resisting, and so-

A paper presented before the Thirteenth Annual Convention of the society held in Boston, September 21 to 25, 1931. The author, H. G. Keshian, a member of the society, is metallurgist, Chase Companies, Inc., Waterbury, Conn. Manuscript received May 6, 1931.

called automobile and aircraft steels, from which they can select the material possessing the right kind and amount of property for which they are looking; but to the user of tool steel such valuable quantitative data is sadly lacking. Terms, such as "keen cutting edge," "toughness," "high wearing property," "freedom from excessive warpage," "resistance to shock or impact in high degree," et cetera, at best are mere conventional phrases to describe certain physical properties. They are devoid of any definite quantitative data. They are not only insufficient to guide the tool designer in his efforts to select proper steel, but also frequently confusing. The present status of selecting tool steel and a necessity for improvement can better be realized by considering the existing methods in general use.

Generally speaking, steels for various tools are selected, at present according to

(1) Tradition,
(2) Personal experience,
(3) Recommendation of steel makers,
(4) Service test.

In the traditional method the mechanic or the engineer chooses the steel according to the established custom or specifications of his plant. In fact, he has no choice in the matter and so far as he is concerned, neither the kind nor the physical properties of the material enter into the picture. Even the reasons for using one type or brand of tool steel in preference to others may not always be clear to him, so a definite knowledge of the necessary physical properties of the steel does not play a determining part in this method of selecting tool steel.

In the second method of selecting tool steel decision is rendered by the personal experience of the engineer or mechanic with various kinds of steels on the market.

It is true that in both these methods the preferential list of tool steels may have been based on so-called service tests. It is commonly known, however, that a steel which may be considered a "top notcher" in one plant does not enjoy even third class position in another plant for the similar class of tools. Furthermore, the same type or brand of steel is quite often differently valued by different engineers or mechanics for similar operations in the same plant. So it is obvious that neither of the two methods just mentioned is satisfactory.

The third method of selecting tool steel is based on the recom-

mendation of the steel maker. He makes the steel and it is presumed he ought to know what characteristic properties his material possesses, therefore, he is the one that is in a position to tell for what purpose his products are best suited. Catalogs published by various steel companies are, therefore, most frequently used. What recommendations do they make to guide the tool engineer or the mechanic in selecting the right steel for a given job?

Perhaps a good way to pursue this inquiry is to review some of the recommendations from the catalogs of several outstanding foreign and domestic steel companies.

Let us assume that we want to select a steel for drawing dies. We find that the Steel Company A says:

"X steel is a water hardening steel of most enduring qualities for tools that must retain their size with a minimum amount of wear, such as drawing and master dies. X steel is one of the longest wearing steels made."

Company B says:

"Steel G is a high grade alloy steel recommended for hard dies for general cold drawing work. It becomes excessively hard with a sort of self-lubricating surface" . . . etc.

Steel Company C says:

"We recommend F steel for blanking and trimming dies subject to very high pressure and severe duty. It will also give excellent service for wire drawing plates for steel wire, copper and brass wire, also for complicated and expensive dies where value is laid upon a keen cutting edge resisting abrasion."

Steel Company D says:

"M, a super die steel that knows no master, designed specially to meet the exacting requirements of accuracy and production demanded by twentieth century efficiency engineers . . . it possesses the maximum resistance to abrasion, sinking and wear of any die steel known."

Steel Company E says:

"R steel is the very highest grade of tempering steel, recommended for drawing dies for brass and copper and for blanking dies for hard materials."

These quotations illustrate the third method of selecting tool steels. They have been quoted here not with a view to criticize the substance of these recommendations. They are of much help to the

person who is in pursuit of the best and carry the weight of confidence and certainty based on the integrity and experience of the maker. Yet they still lack that quantitative information which the engineer is looking for.

If the required chief properties of a header die are resistance to impact and wear, it is reasonable to expect that these properties be given in numerical values in terms of some accepted standard method of testing these properties. Again, if the principal requirements of a cold drawing die are strength and ability to wear, it is quite necessary that these properties of a steel recommended for that purpose be given in numerical values in terms of some accepted method of testing wear and strength so as to enable the user to select the proper steel.

The author in an effort to ascertain the relative values of a few types of tool steels for cold drawing dies made the following experiment.

Eight different brands of steel recommended for drawing dies were selected and from each three drawing dies were made. Then each set of dies was heat treated according to the instructions of the manufacturers. The dies were tried on drawing admiralty tubes under practically similar conditions. A careful record was kept of the number of feet of tubing drawn per die and the amount of wear of the diameter of each die. The commercial rating of the different brands of steel and the wearing properties, as measured in terms of number of feet of tubing drawn per 0.001 inch wear in the diameter of the die, are given in the following table.

In the table the steel costing the most is given a commercial rating of 100 per cent and the steel showing the maximum number of feet of tubing per 0.001 inch wear in the diameter of the die is given a maximum wearing property of 100 per cent.

Steel No.	Commercial Rating Per Cent	Wearing Property Per Cent
1	100	34
2	83	26
3	72	21
4	72	31
5	72	100
6	63	82
7	56	24
8	45	31

From the above table it will be noted that there is no relation between the commercial rating of these steels and their wearing proper-

ties, namely, that the steel that has the highest commercial rating is the third in wearing property and the steel that is fifth in commercial rating is the highest in wearing property.

The fourth method of selecting tool steel is what is called the service test. The basis of this test has a similar meaning in its final analysis as the popular saying that "the proof of the pudding is in the eating"; but one must not forget that there has always been considerable difference in the tastes of the persons eating the pudding.

The service test is perhaps the only test approximating the true test: its value, therefore, should not be underestimated. Yet it is a fact that often the results of the service test greatly differ, in similar applications under presumably similar conditions. Quite often the same kind of steel made into the same class of tools and heat treated similarly but used by different operators show contradictory results. It, therefore, becomes quite evident that even the service test is not to be too confidently considered as a satisfactory method of telling the true and intrinsic values of various kinds of steel. So long as the conditions surrounding such tests often vary from plant to plant, the service test will always be a source of uncertainty and contradiction to the consumer and unfairness to the steel maker. Moreover, the service test often is expensive and lacks the simplicity of a standard test procedure.

If what has been said about the principal methods of selecting tool steel in use at present is true, namely, none of them gives the necessary information regarding the absolute and inherent properties of a given type of tool steel, then we want to know if it is possible to catalog, so to speak, the various properties of tool steels, such as wearing, shock resisting, cutting, shrinking, fatigue and other characteristics, in definite and comparable figures in the same manner that tensile, yield elongation, etc., properties of structural steels are cataloged for the more intelligent use of the structural, automobile and aircraft engineers.

It is realized that the problem is a difficult one. For example, the problem of measuring wearing properties of tool steel is especially so. There is the straight wear of a cold drawing die; there is the impact wear of a header die; or the rolling wear of a ball or roller bearing. How can these different forms of wear be reliably tested or measured? But, even in the fact of these difficulties it must be remembered that man has succeeded in solving even more difficult ones,

and the speed and thoroughness with which he has solved them have depended on the eagerness and determination with which he has engaged his task.

To sum up: The designer of a machine has at his disposal a vast amount of data on the physical properties of many types of steels. For parts of his machine that are subject to wear, tension, compression, impact, etc., he can select steels that fulfill these requirements in various degrees. Similar information is available to the designer of a structure, automobile, vessel, locomotive, dirigible and airplane; but when it comes to tools the designer has only vague and meager information as to the properties of a thousand and one kinds of tool steel on the market, from which he must select his material for tools requiring different characteristics.

Is it not, therefore, worthy of serious consideration of both tool steel maker and the consumer to determine the requisite physical properties of tool steels and make them available to the designer of tools?

DISCUSSION

Written Discussion: By John A. Mathews, vice-president, Crucible Steel Company of America, New York City.

The tool steel business has always been one in which artistry and skill, coupled with intelligent technical selling, brought success. It is not a mere question of men and machines, and quantity rather than quality. It is the "divertissement" of the ferrous industry—pleasure rather than profit. Few industries require so much care, provide such pleasurable variety, offer so much satisfaction or so little hope of great financial reward. The successful tool steel man is imbued with the same spirit of sacrifice as the teacher or missionary. If Mr. Keshian's dream came true that would all be changed and making tool steel would hold all the fascination of rolling merchant bars or assembling bolt and nut number 278 on a Ford.

Second only in fascination to making tool steel is successfully fashioning and treating it. It would seem that anyone with experience in the tailor's trade and ability to read the figures on a tape line could make a good suit. Experience tell us, however, niceties of workmanship and subtleties of style escape some of them and we prefer this tailor to that one. It is the same with tool steel. We have our own preferences and the products are not alike in spite of all the specifications we can write. The subtleties of brands still remain.

Mr. Keshian wishes we could make tools on so-called engineering lines from a quantitative knowledge of various physical qualities which tool steels possess. Is the case quite parallel with that of the automobile or bridge builder? I think not. These engineers have at their disposal great quantities of physical data and engineering formulae. A glance at the properties of the S.A.E. steels

shows great similarity between large groups of steels; by slight changes of carbon or heat treatment identical static properties may be obtained. But did you ever know an automobile engineer who could give a sound metallurgical reason as to why he uses nickel steel while Tom uses chromium-nickel and Harry uses chromium-molybdenum steel for the same job? They have their preferences! Do they rely implicitly on the data or the formulae? They do not, but use a "factor of safety" of three or four. The same is true in building and bridge construction. In the use of large factors of safety their designs differ sharply from those of the tool designer. His stresses are very unequal within a single die and cannot be calculated. Furthermore, the user of the tool or die pushes it to the limit. There is no factor of safety of four. If a high speed lathe tool were worked on this principle it would last almost indefinitely but the production per hour would not satisfy the boss and the operator would be told to "speed her up."

We do not customarily make tensile or impact tests of tool steels, nevertheless, there is a large accumulation of data on these properties of high carbon steels, but little use has been found for it in the practical affairs of tool steel use. Hardness tests are much used and of great help in checking up on heat treatment. Originally, all tool steel booklets asked the buyer to state the purpose for which the steel was to be used and the steel maker recommended the grade and temper. That was a much simpler task twenty-five years ago than it is today. The statements in a catalog have to be of the most general nature, but that is quite different from the specific recommendations of a technically qualified steel representative, after he has been given a chance to study the specific use and the previous experience with the tool.

The special case of steels for drawing dies, mentioned by the author, is interesting but might not a second lot of dies from the same steels have stood in quite a different order? If some of these steels were new to Mr. Keshian and not grades with which he or his organization was experienced, is it fair to assume that each was treated in the very best way for that particular grade? The relative position of some of them might be greatly altered by an extra 25 degrees in the heating or tempering, or five minutes more time. The differences shown between a dozen different steels are not nearly as surprising as differences in performance of, say, a dozen tools off the same bar and with presumably the same treatment, or, stranger still, the life of the four different grooves of a hot bolt gripper die. Suppose we had complete knowledge of tensile strength, impact resistance, hardness, normalcy, grain size, etc., of the steel from which this gripper die was made, would it help us to say why groove one made 16,833 bolts and grooves 2, 3 and 4 made respectively 102,666, 34,766 and 80,723? The chances are that this small die was quite homogeneous and was uniformly heated and quenched by people long experienced in their field. There was no question of difference of design in the four grooves. What chemical, physical or mechanical data could account for this not uncommon result? Furthermore, each groove failed from a separate cause.

Mechanical failures of automobile parts, barring accidents, are becoming rare and duplicate steel buildings do not collapse in a somewhat variable series of years. It seems to be a case of a generous factor of safety and relatively

simple stresses as compared with intricate dies and tools. If tool steel was used in the condition as supplied by the maker there might be some promise of realizing Mr. Keshian's hopes, but when our material of a given set of properties is made into innumerable shapes or designs, transformed by hardening and tempering, used for various purposes in fabricating other materials of variable hardness, erosiveness, toughness, etc., and at variable speeds and with variable stresses in different parts of a single tool, it is hard to imagine that ordinary or extraordinary laboratory data pertaining to the raw steel as furnished can be translated into use value or suitability in the finished tool. Valuable and broad as may be the knowledge and experience of the tool-steel man, yet nothing can equal carefully planned and conducted production tests for a specific job. Having selected a suitable tool steel—either on general reputation, prior experience or upon the recommendation of a qualified expert to whom the problem has been explained—then the influence of variables beyond the control of the steel maker can be studied with profit and it is not fair to a new steel to order only enough to make one tool and then condemn it, perhaps on your errors of design, heat treatment and the hazards of use.

I honestly do not know how to start to solve the problem as outlined by Mr. Keshian. For the present, the tool steel industry bids fair to remain one of personal service and one especially responsive to confidence and cooperation which yield mutually satisfactory results to maker and user.

Written Discussion: By Alvan L. Davis, research engineer, Scovill Mfg. Co., Waterbury, Conn.

Mr. Keshian's plea for definite data of the properties of tool steel may seem Utopian, but the march of progress is in the direction he indicates. Some day his vision may be realized.

It is the properties (obtained by heat treatment) which affect the performance of steel in tools, that Mr. Keshian would see cataloged, evaluated and used as criteria in the selection of tool steel. He lists 5 properties as follows:

"Wearing, shock resisting, cutting, shrinking, and fatigue endurance."

We may consider these properties as follows:

I. Wear resistance, under varying conditions of abrasion and erosion, comprehending wide ranges both of pressure and velocity. This in itself is a large order, but surely is a very attractive field for study.

II. Shock resistance, under single and under repeated blow tests. This can be done, though the ideal testing machine has yet to be invented.

III. Cutting ability. This can be determined by direct test, but could also be estimated (throughout the range of heat treatment) from tables giving both the room temperature hardness and the hardness at red heats; and also from tables giving the coefficient of toughness as determined by the Emmons test.*

IV. Shrinking capacity can be determined without great difficulty, using various standard treatments.

V. Fatigue endurance can be ascertained by methods now in use, for a variety of testing conditions which could be set up as standard.

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*J. V. Emmons, "Some Physical Properties of Hardened Tool Steel," presented before American Society for Testing Materials, June, 1931, meeting.

To the above list of properties we might have to add others such as,

(a) Depth of penetration of hardening under various specified heat treatments.

(b) Permanence of size and hardness after heat treatment during stated periods of time.

(c) Susceptibility of steel to cracking during hardening under standard test conditions.

(d) Machinability of the steel under various conditions of treatment.

(e) Heat treatment required to produce fully spheroidized anneal, and hardness corresponding thereto.

To carry out such a program for even one steel would be a long and costly undertaking. This is what retards development, but progress is being made, first at one point and then at another, as conditions become favorable and economically possible. For example we may cite the valuable paper of J. V. Emmons** in which he develops a method for valuing the coefficient of toughness of a steel for any given heat treatment.

It would seem that we must await the gradual development of such testing methods as justify themselves practically, and the subsequent accumulation of considerable data from them, before we can hope to see the fulfillment of Mr. Keshian's program.

Written Discussion: By J. V. Emmons, metallurgist, Cleveland Twist Drill Company, Cleveland.

This paper by Mr. Keshian states very clearly the problem with which all tool steel users are confronted. The tool steel manufacturer as well as the user is handicapped by the lack of definite numerical values by which the properties of steels may be rated. Some progress is being made in improving the situation. We now have exact methods for chemical analysis by which the ultimate composition of the steel is accurately determined. We have less exact but extremely useful microscopic methods for determining the structural arrangement of the chemical constituents. Accurate methods of determining the useful property of hardness are everywhere available. Methods of determining various other physical properties, have been developed to a greater or less extent, but of these it may be said that they are not yet generally accepted as being criteria of the probable performance of tool steels.

The real problem appears to be to first develop accurate, readily available and low cost methods of determining other physical properties; second to determine the relationship of these properties to actual tool performance and last to determine the requirements of various tool services in terms of the measurable physical properties. The solution of such a problem will at best be a work of years. One promising method of attack would be by some form of joint action supported by the united efforts of the tool steel industry.

Written Discussion: By G. V. Luerssen, metallurgical department, Carpenter Steel Company, Reading, Pa.

No one will disagree with Mr. Keshian as to the necessity of more informa-

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**See footnote on page 240.

tion on the fundamental properties of tool steels. However, the user has not been alone in sensing the need of such information. If the user has been handicapped in the intelligent application of tool steels, the maker has been equally handicapped in designing them. This fact results from the difficulty in obtaining consistent data from the only recognized type of test available, namely, the service test, due to wide variations in tool design, method of tool making, physical properties of material being worked and shop practice in the various plants.

The problem proposed by Mr. Keshian presents two phases, first, the development of methods of testing to determine the fundamental properties desired, and second, the intelligent application of these properties to the designing of the tool. Both phases present serious difficulties. To take up some of the various properties enumerated by Mr. Keshian, comprehensive wear data, for instance, are extremely difficult to obtain. Figures which apply to wear against brass do not necessarily apply to wear against steel or other metals. Those which apply when one lubricant is used do not apply in connection with another. Pressures and surface preparation are still other factors which necessarily have to be considered.

Shock tests of the notch impact type are practically useless on extremely hard steels, and a different type of impact test, one in which the stresses are less localized will therefore have to be devised. While such a test can and undoubtedly will be developed, it will have to be carefully interpreted in connection with some of the tool steels such as for instance the relatively shallow hardening types. Surface preparation of the specimens will have to be carefully standardized.

Tests for shrinkage or size change can readily be made, but not so easily standardized, particularly where shallow penetration steels are involved, in which the profound effect of hardness penetration on size change is well known. Penetration tests upon such steels to supplement the shrinkage tests would undoubtedly be of assistance.

Assuming now that test data such as outlined above were available, the application of such data to tool design would still present a major problem. This problem would likely have to start with standardization of tool making and tool hardening methods. For instance it is well known that the practice of cold hubbing die impressions has a profound effect on resistance of the steel to fatigue and impact under certain conditions of service. Likewise surface finish has a direct effect upon such properties. Slight faults in heat treatment and in grinding also have their influence. Such variables would therefore have to be eliminated or at least controlled so that the condition of steel in the ultimate tool would compare within reasonable limits with the ideal condition obtaining under the fundamental test conditions. The physical and micro-properties of the material being worked likewise have an important effect on tool performance, and attention should be given the control of these factors.

Having satisfactorily disposed of such variables, quantitative data such as suggested by Mr. Keshian would undoubtedly be of assistance to the tool designer even though they did little more than place the various steels available in their proper sequence as regards the several properties discussed. The application of these properties in the design of the tool would then require consider-

able skill on the part of the designer. Careful compromises would probably be necessary at times in order to secure the proper balance of properties. Even though the tool were designed under such conditions, occasional check-ups on service results would be desirable, since service after all is the ultimate criterion.

Mr. Keshian opens a fertile field of thought in suggesting that rule-of-thumb methods be eliminated from the selection of tool steels. The writer has attempted to point out some of the possible obstacles in such a course, which though bothersome are not necessarily insurmountable. As Mr. Keshian points out in his paper, greater problems than these have been solved, and it is hoped that the necessary eagerness and determination mentioned by Mr. Keshian will not be lacking in the present case.

Oral Discussion

A. H. D'ARCAMBAL:[1] All large users of tool steel will undoubtedly agree with the recommendations so well discussed in Mr. Keshian's paper. Unfortunately we do not have suitable machines or instruments at the present time that can be used for determining all of the physical properties of tool steel that the author believes should be available to the user. The Amsler machine is used to some extent for determining the wear-resistance of metals but the results obtained are seldom borne out in actual practice. As an example, Mr. S. J. Rosenberg in his excellent paper on "Resistance to Wear of Carbon Steels" finds that normalized high carbon tool steel with a Rockwell hardness of B-105 shows almost as great resistance to wear as hardened carbon tool steel with a Rockwell hardness of C-60. The special wear testing machine built at the Bureau of Standards a few years ago gave results that were far different than actual tests indicated.

It has always been a problem to devise some efficient means of measuring the toughness of hardened tool steels. Tensile, transverse and impact tests have given extremely erratic results. Since reading Mr. Emmons' paper on "Some Physical Properties of High Speed Steel," we are convinced, however, that carefully conducted torsional tests will furnish us with valuable information as to the toughness of hardened tool steels of different compositions and subjected to various hardening treatments.

In regard to the effect of various compositions on the size change taking place during the hardening operation, we believe that data of this type could easily be obtained through carefully conducted tests by the metallurgical department of the mill, using specimens of different shapes and sizes. We have learned through experiments in our plant that carbon steel taps of a certain size made from one type of steel will lengthen in hardening whereas taps of the same size and type made from a different composition will shorten. When a sales representative introduces a new type of tool steel, he is usually not in a position to state the size change to be expected from this improved type of steel, being merely informed by his company that this material should make better tools.

Today the tool steel salesman must contact with the metallurgist in all large metal working plants. The sales representative cannot at times answer questions as to the analysis of a certain type of steel, machining qualities, proper

[1]Metallurgist, Pratt and Whitney Company, Hartford.

hardening treatment with resultant hardness, etc., due to the mill not keeping their sales force properly informed regarding their product. We sincerely believe that both the manufacturer and user would benefit through having the tool steel salesman so equipped as to be able to intelligently answer reasonable questions regarding his product.

The quality of tool steel of American manufacture today is superior to that of a decade ago. It is the wish of the writer that 10 years hence, the mills in this country will be in a position to furnish us with the physical properties of tool steels along the lines suggested by Mr. Keshian, resulting, undoubtedly, in a superior product due to increased knowledge.

S. J. ROSENBERG:[2] I would like to add a word relative to what Mr. d'Arcambal has said in regard to the wear test on tool steels. It is difficult, as we all know, to make wear tests which are indicative of what we obtain in service. A wear test, to be of any use, must be so designed that it actually simulates the conditions which we obtain in service. If we wish to make a wear test upon tool steel, the test should be of such a nature as to approximate the conditions we would obtain in practice. The tests to which Mr. d'Arcambal referred in which we obtained rather different and apparently contradictory results with a normalized and hardened high-carbon steel were entirely different in nature. In one case the wear tests were made under rolling friction accompanied by very high pressures and in the other case the tests were made under relatively light pressures and under sliding friction. We cannot expect to get the same results in both types of tests, and that is the reason why in service two steels will rarely, if ever, show the same relative efficiencies when subjected to different kinds of service. The fact that there are many varieties of wear involving so many complex factors makes wear testing a difficult problem.

If the steels tested by Mr. Keshian had been subjected to some other type of test than the one devised by him, he would probably have secured results entirely different from the results reported by him. This one point cannot be too greatly emphasized—results secured by two different methods of wear testing should not be expected to be similar and from this it follows that results secured from any one method of wear testing should be applied only to those particular conditions of service which that wear test simulates.

Author's Closure

The idea of expressing the more useful properties of tool steels in numerical values is not altogether new; but it is new, at least as an attempt to bring it forward as a consumer sees it. Its adoption, like many new ideas, is bound to disturb to a certain extent some of the existing conditions of handling tool steel. It is to be expected, therefore, that at first glance it will appear somewhat Utopian. It is gratifying, however, to note from the foregoing discussions that the usefulness of the idea is generally conceded and its adoption advocated.

I am glad that Mr. Davis with his wide experience both in the manufacture and application of tool steels agrees with me on the necessity of studying and expressing the physical properties in a manner most useful to the consumer. As Mr. Davis states, there is yet a considerable amount of work to be done be-

[2]Associate metallurgist, U. S. Bureau of Standards, Washington, D. C.

fore we can have sufficiently reliable data; but it is obvious that if we are once thoroughly convinced of the economic value of such data any difficulty—technical or otherwise—will surely be overcome.

Mr. Emmons has already started a series of valuable investigations in this direction. His suggestion that with the joint action of tool steel industry, low cost methods to determine the properties of tool steels be developed and the relationship of these properties to actual tool performance be determined, is certainly to the point. It is hoped that his suggestion will be given serious consideration.

Mr. Rosenberg's statement that the testing methods in order to be of practical value should approach the working conditions as nearly as possible, is important. The lack of the full appreciation of this single point alone has been one of the main contributing factors to the failure of many attempts to devise methods to satisfactorily test the properties of tool steels.

Mr. d'Arcambal, likewise, realizes the difficulty of devising dependable testing methods. We must, of course, go a step further and make systematic and serious attempts to overcome these difficulties as I have stated at the conclusion of the paper. Mr. d'Arcambal suggests that the tool steel mills supply their representatives with more definite knowledge on the properties of their product. It is to be hoped that his suggestion will be heeded and that his prediction that such data will be available 10 years hence will become a reality.

Mr. Luerssen very clearly grasps the problem confronting both the maker and the user. He divides it into first devising methods to test the properties of tool steels, second, to intelligently apply the data to practice. He sees considerable difficulty to be overcome in either case.

I am glad that he agrees with me on the necessity of this information. In regard to the difficulties, both in securing data and in applying them, I must repeat the same statement which I made to some of the other gentlemen, namely, that the first requisite is to truly appreciate the value of this information, then the difficulties whether in devising proper testing methods or in applying the results to practice will not appear as great as they do.

Dr. Mathews gives us the other side of the picture, namely, that of the manufacturer. As contrasted to Mr. Luerssen's discussion, his is even more conservative. His views, however, as coming from an illustrious leader in tool steel industry are of particular value and worthy of serious consideration.

Whenever Dr. Mathews has discussed the subject of tool steels he has not failed to stress the artistry and skill involved in making them, charmingly painting, in his lucid manner, the skill, devotion, and personal pride to be found in the trade. Thus he idealizes the quality as something intangible, not subject to measurement or definition by any kind of specifications.

We must, of course, appreciate the skill and the artistry necessary for the production of fine tool steels; but when we are confronted with the serious problem of selecting the right kind of steel for specific application, then the mere acceptance of artistry and skill cannot be of much help and it becomes necessary to know also the particular characteristics of various steels in measurable form. Dr. Mathews not only predicts an adverse effect of such

information on the tool steel industry, he also doubts the usefulness of this information to the consumer. Let us consider these two objections separately.

Dr. Mathews thinks that if we were to rate the tool steels according to their physical properties, like S.A.E. steels, the manufacture of tool steels would lose its fascination and would become an uninteresting routine.

Let us assume that this will happen and if it happens, I assume that the worst possible effect on tool steel industry would be to lower the quality of tool steels. If tool steel manufacturing loses its fascination would it lower the quality? Not necessarily. To take a familiar example, consider the process of heat treating steel. This, from the time of Homer until recently was held to be an art requiring great personal pride and devotion on the part of those who practiced it and was kept as secret as possible. When the modern heat treating equipment began to replace the blacksmith's forge and various temperature controlling devices started their forward march to the heat treating department, the loss of interest on the part of the heat treater was then likewise anticipated and lamented; but look! what has happened. Could we do with the fascinating art of heat treating of the past what we are now able to accomplish with rather routine method of heat treating of today? Are the products of heat treating now inferior to those of the past? Certainly not. The pleasure of doing things is not the only incentive to quality, not even the greatest. That place must be given to what we may tersely term legitimate profit, without which it is difficult to imagine that many would produce for the pleasure of it. We are all anxious to have people beat a pathway to our cabin, but in order to get them to do that we must make the proverbial mouse trap of better quality whether we like the drudgery of doing it or not.

Dr. Mathews refers to another phase of the subject of tool steels which I purposely wanted to avoid, namely, the question of brands. He points to the futility of trying to evaluate the properties of tool steels by specifications, saying that "the subtleties of brands still remain." This is a fertile subject for discussion and however tempting I was still going to try to avoid it lest I deviate too far from the main purpose of my theme, but inasmuch as Dr. Mathews has graciously drawn me into this technical entanglement I would like to take up just that part which directly bears on my paper, that is, the question of measurability or unmeasurability of the properties of brands.

It must be admitted that one brand of tool steel may have different properties than another brand of identical chemical analysis. Yet it is difficult to see why these properties which give rise to this difference cannot be measured. It is said that there are more than 600 brands of tool steels on the market. If this is so, each of these brands must have its own subtleties. Now, if these subtleties could be so controlled, manipulated and compounded as to produce so many brands, then it follows that these subtleties must have been capable of measurement; because, if the manufacturer cannot measure these subtleties either directly or indirectly, then he cannot control them, and if he cannot control them, he surely cannot produce so many distinctly different brands and duplicate them day after day. Whether the manufacturer actually does this or not is a different matter in which we are not interested here, but what we are interested in here is that when we say that the properties or the subtleties

of brands cannot be measured we are throwing overboard the very thing upon which the idea of brands must rest. In other words, if we cannot measure properties we cannot produce brands.

To the suggestion that the properties of tool steels be available to the tool designer in the same manner that the properties of structural steels are available to the structural engineer, Dr. Mathews does not think that the two cases are similar, saying that while the structural engineer uses a factor of safety, the tool engineer has no factor of safety and the user of the tool or the die pushes it to the limit, from which he concludes that it would not do the tool designer much good to know the properties of the material he uses. But I believe that the tool designer should have even more data on the properties of the material he uses, than the structural or automobile engineer.

The automobile or structural engineer has usually known stresses to meet, and having before him a variety of steels of known properties he can use the steel he wants by using a certain factor of safety. The tool engineer on the other hand may have conditions where he can tell the type and the magnitude of stresses he must meet and he may have other conditions where the stresses will not be so well defined. Now, let us consider these two sets of conditions separately and see of what assistance a more definite knowledge of the properties of tool steels would be to the tool engineer.

Where the stresses are known the tool engineer can follow the similar procedure as the structural engineer using proper safety factors; but how can he do that without the necessary data on these materials?

Under the second set of conditions, that is, where the stresses are not well known, he must resort to the cut and try method. In other words, he must follow either one or all of the four methods of selecting tool steel previously described in this paper, which, briefly, means that he must try as many kinds of steels as possible until he strikes the one that will meet his requirements; but until then, oh! what trial and tribulations he must go through, often duplicating considerable time and energy-wasting-work and frequently stopping where he had started first! If he had at his disposal the data that we have been discussing, the solution of his problems would thus be much simpler. He could, first, size up the approximate type and the magnitude of the stresses in any particular case, then scan the various properties of different types of steels available and select those steels that would most likely meet his conditions and then, only then, proceed with his cut and try method. But how can he do this without some reliable data?

Dr. Mathews says that the information given in catalogs must be of a general nature. Whatever the reason or necessity for making these catalogs of a general nature, they are certainly too general for the purpose they are used by the customer. It is, however, gratifying to note that some of the tool steel catalogs are now coming out with a little more specific information.

Regarding the test results of the eight different kinds of steels given in the paper, Dr. Mathews says that possibly I or my organization was not familiar with some of the steels, that some changes in their heat treatment might have reversed the results. I might say that none of the instructions from the steel makers for heat treating their steels was out of the ordinary as to make the

heat treatment of these steels too complicated or render their handling something outside of our many years of experience of hardening dies. Despite this fact, I am willing to admit that possibly we did not handle each steel in the very manner that it should have been handled and, as Dr. Mathews says, maybe that using a hardening temperature of 25 degrees higher or lower or soaking the steel five minutes longer at heat would have also changed the results; but if a steel cannot be handled within 25 degrees variation in heat treatment and five to 10 minutes variation in soaking time, that steel can hardly be considered a commercially usable material for this class of service.

The results were the averages of three separate hardenings and shrinkings representing nine tests for each steel or 72 tests for the eight steels. If there was a possibility of incorrect handling some steels, the same possibility of making errors existed for the other steels. Furthermore, the variations in wearing properties are so large that it is difficult to explain them on the basis of slight errors that might be made in heat treatment.

We now come to another point in Dr. Mathews' discussion, namely, that if we had all the data on the various properties of tool steels, would we then be able to explain the failures and irregularities at times noticed in the performance of tools made from the same bar of steel and heat treated, presumably in the same manner? I do not claim that these data will explain all such failures and irregularities; but the fact that they cannot does not prove that they are of no assistance to the consumer; because there are other factors besides the intrinsic properties of tool steel which enter into the performance of tools. Yet such data will indirectly aid in the explanation of such failures and irregularities. For instance, if a set of tools were made from a steel of known properties and the tools did not function properly, then it is up to the men up and down the line, from the designer to heat treater and the operator to look for an explanation and carefully go over their respective methods of handling the tool. This procedure gives the steel a bill of health, so to speak, and eliminates one of the main variables, namely, the question of right steel and thus greatly simplifies the solution of the problem. As contrasted to this take the case where the type of steel used and the method of handling are both in doubt, then the explanation of these failures becomes increasingly difficult and often it becomes a draw between the maker and the user.

No mention was made in the paper of the expert service from the tool steel mill which is always available to the consumer in the selection of the proper steel for a given purpose which, as Dr. Mathews states, is a bond of confidence and cooperation between the two parties. This is, indeed, a very important feature of the tool steel industry and perhaps will always be so. It must be remembered that this service, valuable as it is, is of occasional nature and not permanent and systematized data upon which the tool designer can lay his hand whenever he needs them to guide him in his daily work of selecting steels, an ever-ready guidance which is necessary, particularly in large industrial plants where the problem of selecting steels comes up often with bewildering variety and rapidity almost every hour of the day. So it is quite obvious that along with this valuable expert service we must also have easily and quickly available data on the various properties of tool steels.

COMPARATIVE TESTS OF HARDENED STEEL BY MEANS OF PENETRATION TESTING APPARATUS

By N. N. Sawin and E. Stachrowski

Abstract

Three methods of determining the hardness of hardened steels are described, their suitability is discussed and hardness conversion curves given. It was found that:

(1) The Vickers machine is well adapted for individual and laboratory tests of hardened steel, especially if shape is intricate.

(2) The Rockwell machine has its place when work of simple shape is being tested on a production basis.

(3) The Monotron machine may be of advantage when tests on a production basis have to be performed and at the same time the state of the surface has to be investigated.

(4) Brinell hardness numbers up to 500 and Monotron numbers HM C-D show a linear relation to Vickers numbers.

(5) Rockwell C numbers from H_{RC} 20 to 50 may be converted to Brinell, Vickers and Monotron numbers.

Comparison of Methods

THE Brinell method of testing hardness by the indentation of a steel ball is not suitable for hardened steel, because of the flattening of the ball during penetration wherein a permanent deformation takes place. G. W. Quick and L. Jordan[1] have closely investigated this fact and have suggested the manufacture of balls from an alloy steel Fe-C-V by cold working, such balls showing the least deformation. However, the reduction of permanent deformation does not render the balls any more suitable for testing hardened steel, because permanent (plastic) deformations are always accom-

[1]G. W. Quick and L. Jordan, "Iron-Carbon-Vanadium Alloy for Brinell Balls," Transactions, American Society for Steel Treating, Vol. XII, July, 1927, p. 3.

The author, Prof. N. N. Sawin, is chief of the toolroom department of the Limited Company, formerly the Skoda Works, Plzen Czechoslovakia; Mr. E. Stachrowski is engineer of tests in the same department.

panied by elastic ones which upset the results, too. The hardness of the tested balls ranged between C64.5-C65 Rockwell. According to our tests on a Vickers diamond hardness testing machine steel balls manufactured according to Hultgren have a Brinell hardness of 952, which is of about the same hardness as properly hardened alloy steel.

To investigate the influence of the steel ball on Brinell hardness tests Shore[2] and Mailander[3] used ball-pointed diamonds having a diameter of 10 and 5 millimeters. The latter came to the conclusion that work-hardened balls will produce a correct impression only when the Brinell hardness is below 500. Shore claims that errors in measuring the impression of a 10-millimeter ball in hardened steel may produce differences amounting to 75 Brinell numbers. This reason, along with many others, as for instance: large, visible impressions which often mar the work, high pressure and consequent impossibility to test thin work, grinding of the test surface, tedious adjustments and reading of the microscope, inability to test work of a complicated shape, et cetera, was responsible for the gradual abandonment of the steel ball and adoption of the diamond for testing hardened steel. A diamond requires smaller dimensions and less pressure. Firth, Sheffield in his "Hardometer" uses a ball-pointed diamond with a pressure of 120 kilograms. The machine is used the same way as the Brinell machine. Its advantages are, (1) Simplicity of operation, (2) Smaller weight than the Brinell machine, (3) Quick adjustment and reading of the microscope, (4) Smaller impression, (5) Possibility to test thin work and complicated shapes. Its disadvantages are, (1) Grinding of the surface to be tested, (2) Loading by means of a spring, (3) Light required for the microscope.

The same method is used in the Vickers diamond hardness testing machine; the diamond, however, has the shape of a pyramid with an apex angle of 136 degrees and the load is generally 50 kilograms.[4] The impression has the shape of a square, the diagonal of which is measured with an accuracy of 1 μ and by means of tables the results are converted into Brinell numbers. The machine has the

[2]A. F. Shore, "Report on Hardness Testing," *Journal*, British Iron and Steel Institute, Vol. 98, 1918, p. 59.

[3]R. Mailander, "Die Härteprüfung von gehärteten Stählen," *Stahl und Eisen*, Vol. 45, 1925, p. 1769.

[4]N. N. Sawin, "Les Résultats Comparatifs des Diverses Méthodes de Mésure de la Dureté des Métaux Trempés, *Génie Civil*, Vol. 89, 1926, p. 159.

same advantages and disadvantages as the "Hardometer," the impression, however, is much smaller (0.3-0.6 millimeter across) and on account of its sharp definition may be measured exactly. Also impressions of different magnitude are geometrically similar and on that account subject to the law of similarity. With the same apex

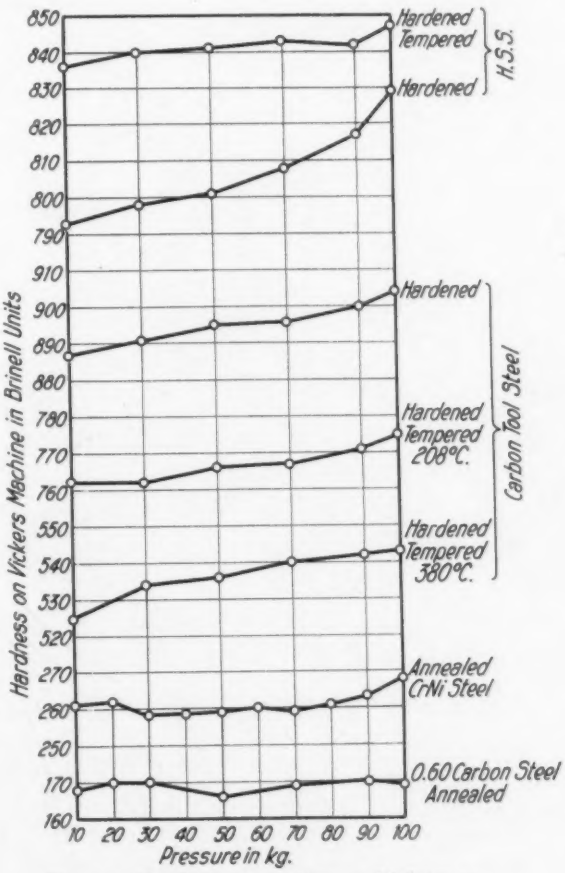

Fig. 1—Hardness According to Vickers at Various Pressures on Several Different Materials.

angle and varying loads up to 100 kilograms almost identical hardness numbers are obtained. R. L. Smith and G. E. Sanland[5] found that pyramidal diamonds with an apex angle of 136 degrees give hardness numbers approximately equal to numbers obtained with a steel ball with a diameter of 1 millimeter under a load of 30 kilograms. The laboratory of the Skoda Works toolroom, being the owner of a Vickers diamond hardness testing machine for the last five years, has determined the hardness of various steels on this

[5] *Stahl und Eisen*, Vol. 45, 1925, p. 1442.

machine with a pressure of 10, 20... up to 100 kilograms. The results are plotted in Fig. 1.

Annealed 0.60 per cent carbon steels and chromium-nickel steels show a variation in hardness within the limits of manufacturing deviations and within the limits due to observation errors. Hardened steels show a regular increase in hardness with increasing pressure (and penetration), most likely because the core is harder than the

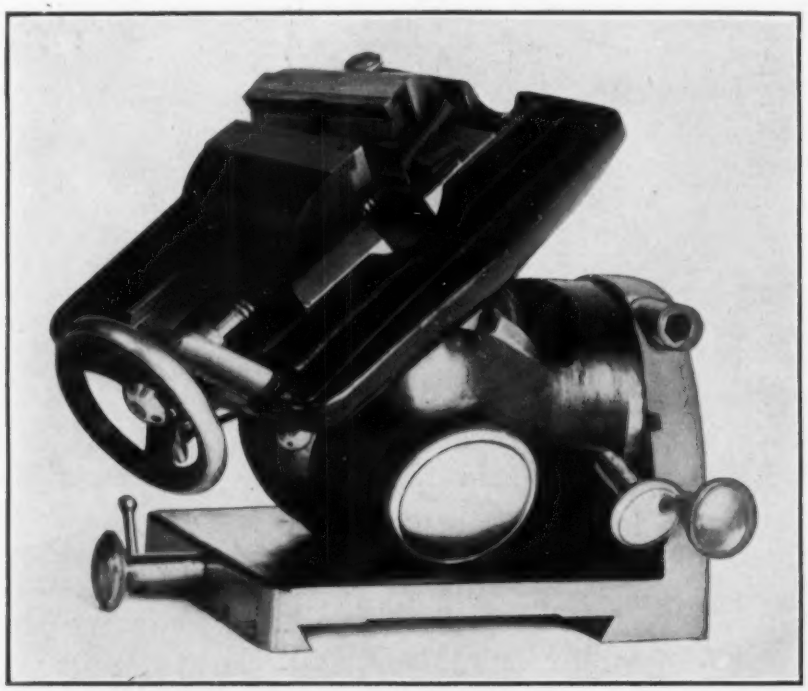

Fig. 2—Showing the Special Vise that was Designed for Testing Complicated Shaped Pieces.

surface. On that account it is better to mention the load when stating the hardness. Due to the relatively small load and to the possibility of measuring the impression by means of a microscope, the tests may be carried out close to the cutting edge, even if the shape should be rather complicated. To this end we have provided the machine with a special vise (Figs. 2 and 3), by means of which the surface to be tested may be levelled up and placed under the microscope. This vise increases the usefulness of the machine to a considerable degree. A test on this machine requires a little more time and the surface to be tested must be clean.

Other methods for testing hardened steel are based on the determination of the penetration depth while making the test. One

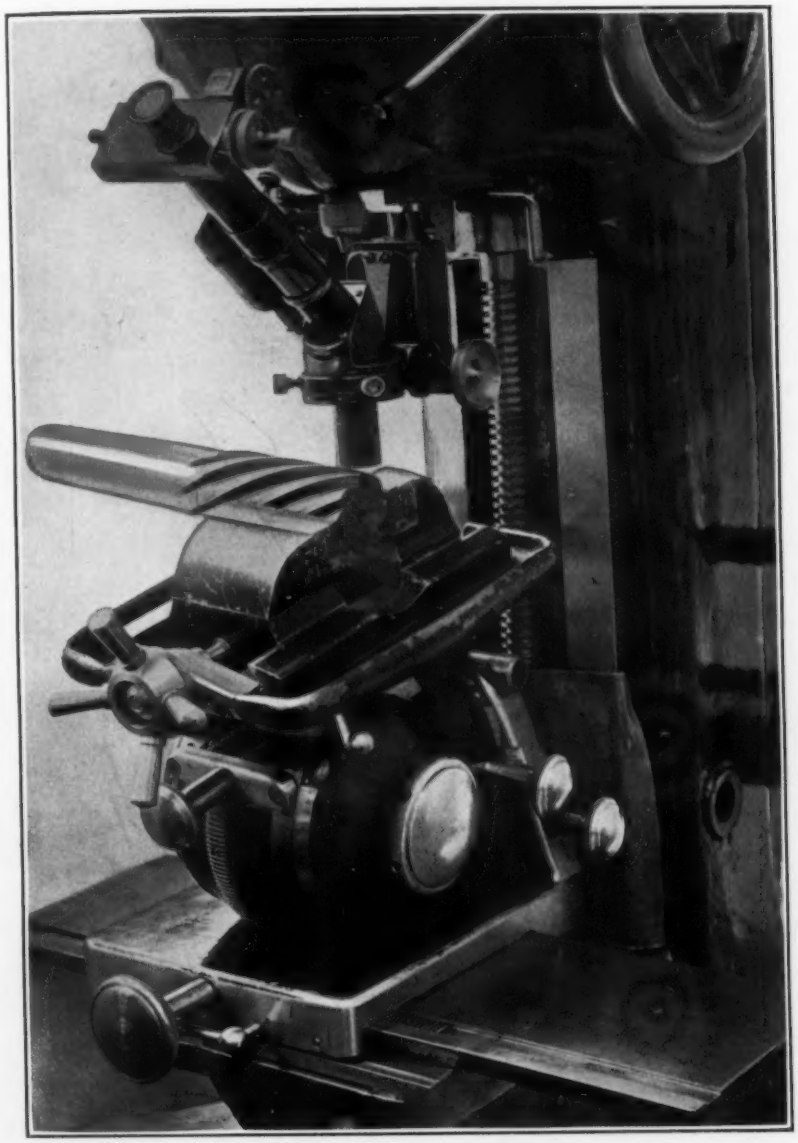

Fig. 3—Photograph Showing the Special Vise in Position on the Vickers Machine Holding a Reamer.

of them, the Rockwell method, uses a cone-shaped diamond with 120-degree apex angle under a load of 140 kilograms. As it is well known the diamond is under a load of 10 kilograms before and after the penetration to insure a contact with the test piece. Work with this machine is simple and rapid. A test takes 15 to 20 seconds, while in the Vickers machine the penetration requires exactly 30 seconds, and the test about 90 seconds, not counting the cleaning of the test

surface. However, on the Rockwell apparatus only work with parallel faces or round work may be tested, as it must not be fastened. It happens quite often that the lower face of the test piece does not seat properly or that it is springy, or it may be covered with scale. During penetration it will give and the apparatus will indicate a lower hardness. For that reason the manufacturers[6] of the apparatus recommend cleaning the seating face of the test piece or making an additional test. This of course also means an additional loss of time. Generally the additional test is chosen and made in a spot close to the first test and a higher hardness number is obtained.

From the foregoing it is obvious that the Vickers machine is more suitable to laboratory work and to occasional checking tests, while the Rockwell apparatus is well adapted for routine tests in production work.

Another machine in which the penetration depth of a ball-pointed diamond (diameter 0.75) is measured, is the "Monotron," invented by Shore.[7] This method is similar to the method of Martens-Heyn.[8] The Monotron method, unlike the Rockwell method, is based on constant penetration (9/1000-inch), the depth of which is measured during penetration, thereby measuring not only the permanent but also the elastic deformation of the impression. The influence of the deformation of the penetrator mechanism and of the test piece on the depth of penetration is eliminated by measuring the penetration depth from the surface of the test piece.

In the Rockwell machine the penetration depth is measured from its body. Determination of hardness by the "constant diameter" method (C-D hardness on the "Monotron") is also very simple and rapid. The test requires 15 to 20 seconds. The hardness is expressed in kilograms of pressure and therefore is proportional to hardness determined by the Brinell method. However, the carrying out of the test as to speed of penetration and reading the depth and load gage, depends very much on the operator. On that account the hardness numbers are subject to much dispersion; for instance on a piece of hardened and tempered high speed steel 20 x 20 x 80

[6]Rockwell Tester Instruction Book, p. W-M 8d.

[7]A. F. Shore, "New Methods in Measuring Hardness," *Heat Treating and Forging*, May, 1929.

[8]P. W. Döhmer, "Die Brinellsche Kugeldruckprobe," p. 32.

the Rockwell machine showed C63.3 to C64.2, that is a dispersion of 1.4 per cent; the Vickers machine 826 to 851 (3.0 per cent) and the Monotron 90.0 to 94.0 C-D (4.5 per cent). The results on the Monotron are greatly influenced by the smoothness of the test surface, freedom from scale, annealing due to improper sharpening[9], work hardening, etc., as the depth of penetration is relatively small. For this reason testing by means of the Monotron discloses mainly the character of the surface which is of importance in automobile parts, tools, etc. The usefulness of the Monotron could be very much extended by providing it with a universal vise. By plotting the loads corresponding to different depths, the progressive hardness from the case to the core may be determined, the depth of case-hardening, decarburization, etc. Altogether this apparatus offers new possibilities of investigation, which may yield in time valuable results.

Conversion of Hardness Numbers

In actual practice the hardness numbers according to one method must be often converted into other numbers. To this end the Skoda Works have conducted a series of comparative tests, the results of which are plotted in chart Fig. 4. The hardness tests were performed on checking bars furnished with the several machines and on other bars of chromium-nickel steel, special tool steel and high speed steel. Hardened carbon steel was not used in these tests, mainly to avoid the fouling of the tests by the considerable difference of hardness between the surface and the core, which was already pointed out by Heller[10]. In addition to these Monotron tests were plotted to investigate the hardness at different amounts of penetration.

As a basis the Vickers hardness numbers were taken at a load of 50 kilograms because it is expressed in Brinell units, generally in use and applicable to hardened steel, too. Three years ago we checked the Vickers hardness on 5 prisms and between these tests and those conducted now we have obtained a difference of 5 points maximum only on a hardness of 918, that is 0.6 per cent.

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[9]N. N. Sawin, "Influence of Regrinding Practice," *American Machinist*, Vol. 70, p. 505.

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[10]A. Heller, "Conversion of Hardness Numbers," *American Machinist*, Vol. 70, 1929, p. 525.

The Vickers hardness was compared with: Brinell hardness; C-D Monotron hardness; C Rockwell hardness.

The Brinell hardness was determined on the "Alfa" machine with a work-hardened Hultgren steel ball having a diameter of 10 millimeters and a load of 3000 kilograms. It shall be designated by $H_{B10/3000}$. According to our tests this hardness is always smaller than Vickers hardness designated by H_{V50}. Hardened steels with 500 H_{V50} and over show a rapidly increasing difference due to permanent and elastic deformation of the steel ball[1]. Below 500H_{V50} this difference amounts to 3 to 4 per cent, which is within the limits of material variations.

Our previous tests[4] with 1 millimeter balls as well as Hankins[11] with a 136-degree pyramid (both on the Vickers machine) have shown that the hardness numbers $H_{B1/30}$ and H_{V50} are greater than $H_{B\ 10/3000}$ by 6-11 per cent.

Brumfield[12] has found on a Rockwell machine with steel balls $\frac{1}{16}$ and $\frac{1}{8}$ inch in diameter and with a 120-degree cone-shaped diamond by measuring the impression, that the Brinell hardness calculated from them is considerably greater than the hardness determined by using a 10-millimeter steel ball. This has been verified by Petrenko[13], who, however, has calculated the Brinell hardness from the penetration depth of balls and diamonds. He explains this difference by the influence of the elastic deformation of the test piece and of the steel ball, which at small loads and with small penetrators diminish the impression produced to a greater extent than when a steel ball of 10 millimeters and a load of 3000 kilograms is used. However, this pheomenon undergoes a further

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[1]G. W. Quick and L. Jordan, "Iron-Carbon-Vanadium Alloy for Brinell Balls," TRANSACTIONS, American Society for Steel Treating, Vol. XII, July, 1927, p. 3.

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[4]N. N. Sawin, "Les Résultats Comparatifs des Diverses Méthodes de Mésure de la Dureté des Métaux Trempés, *Génie Civil*, Vol. 89, 1926, p. 159.

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[11]A. G. Hankins, "Hardness Tests Research," *Machinery*, London, 1926, No. 742, p. 424.

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[12]R. C. Brumfield, "Comparison of Rockwell and Brinell Hardness," TRANSACTIONS, American Society for Steel Treating, Vol. IX, No. 6, p. 841-851.

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[13]S. N. Petrenko, "Relationships Between the Rockwell and Brinell Numbers," Bureau of Standards, *Journal of Research*, July, 1930, p. 19.

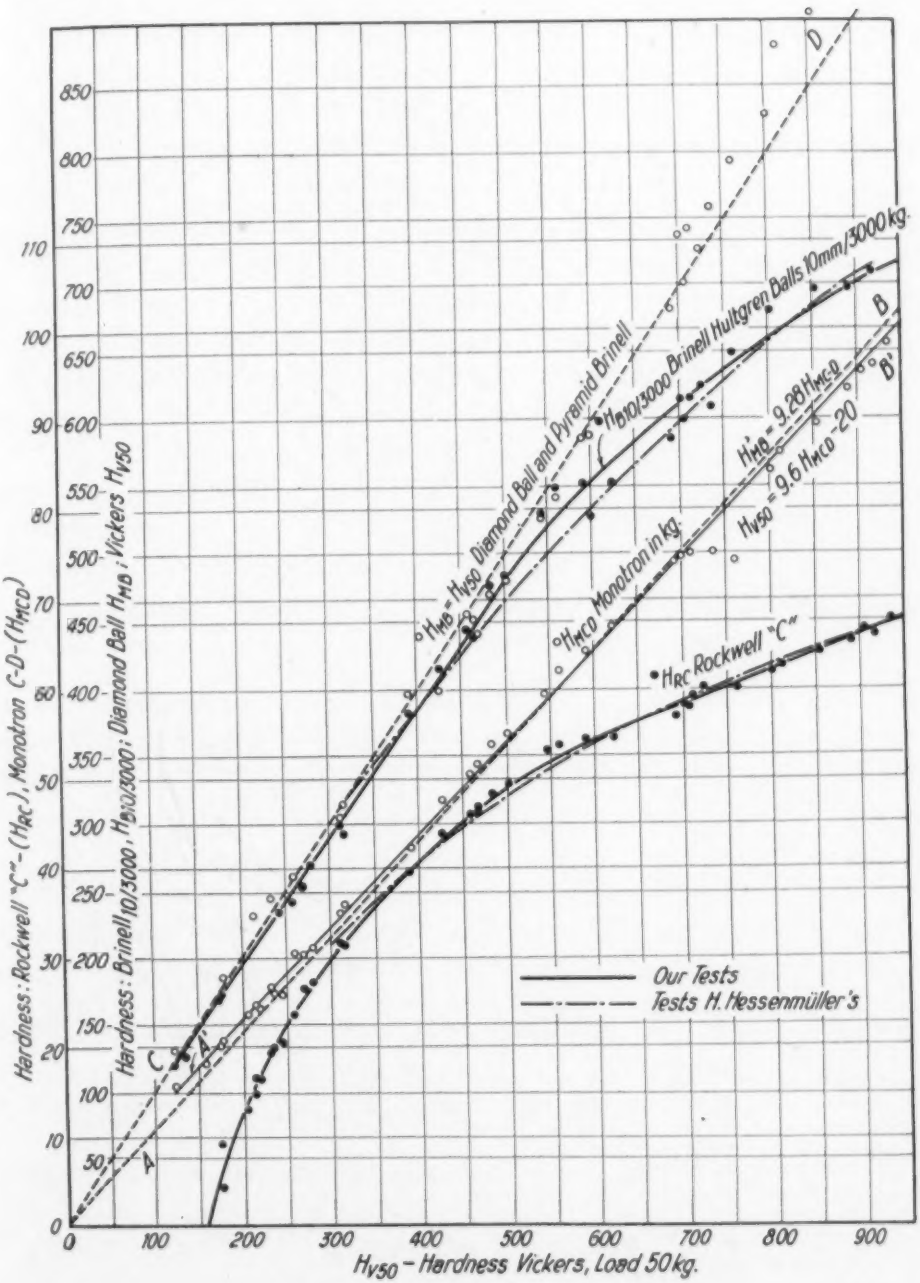

Fig. 4—Chart Comparing Hardness Values Obtained on Several Hardness Testers Compared with Vickers Hardness Numbers Obtained on the Same Test Pieces.

complication, because in the Brinell test we have to also deal with plastic and elastic deformations, as well as with work hardening under the steel ball (according to Brumfield up to H_{RC}=60) and

the forming of a "flash" around the impression. The Vickers hardness is like the Monotron C-D hardness, more of a surface test. The depth of penetration varies from 0.045 to 0.115 millimeter, while in the Brinell test the penetration depth for the same range of hardness amounts to from 0.128 to 0.670 millimeter. These various facts explain the smaller difference in hardness, which, however, is of no practical importance when Brinell tests have to be substituted by Vickers tests. In testing soft steels care, however, must be taken to remove with a file any layers which might be work-hardened by some mechanical machining operation. In great hardness numbers we notice a greater dispersion of Brinell hardness on account of the unsuitability of this test to hardened steels. This may also explain the considerable difference between our curve and that of Hessenmüller[14] in his article: "On the Determination of Surface Hardness and Depth of Carburization of Hardened Pieces." Nevertheless, we see a considerable agreement of our curves of hardness H_{RC} and hardness H_{V50}. The Brinell hardness and the "C" Rockwell hardness have been compared frequently and many articles have been published regarding it. We are quoting them in chronological order; S. C. Spalding[15], I. H. Cowdrey[16], R. C. Brumfield[12], S. N. Petrenko[13], R. R. Moore[17], Wallichs-Schallbroch[18], A. Heller[19].

The Wilson-Maeulen Co. furnishes also a comparative table with the Rockwell apparatus. The conversion curves established by different authors are not quoted in this article, as they would make

[14]H. Hessenmüller, "Bestimmung der Oberflächenhärte," *Werkzeugmaschine*, 1930, No. 3, p. 41.

[15]S. C. Spalding, "Comparison of Brinell and Rockwell Hardness," *American Machinist*, Vol. 61, p. 759.

[16]I. H. Cowdrey, "Relation Between Rockwell and Brinell Hardness Scales," TRANSACTIONS, American Society for Steel Treating, Vol. VII, Feb., 1925, p. 244.

[12]R. C. Brumfield, "Comparison of Rockwell and Brinell Hardness," TRANSACTIONS, TIONS, American Society for Steel Treating, Vol. VII, Feb., 1925, p. 244.

[13]S. N. Petrenko, "Relationships Between the Rockwell and Brinell Numbers," Bureau of Standards, *Journal of Research*, July, 1930, p. 19.

[17]R. R. Moore, "Relationships Between Rockwell, Brinell and Scleroscope Numbers," TRANSACTIONS, American Society for Steel Treating, Vol. XII, 1927, No. 6, p. 968.

[18]A. Wallichs and A. Schallbroch, "Umrechnung von Härteziffern bei Stählen und Einigen Metallen," *Maschinenbau*, Vol. 8, 1929, No. 24, p. 824-827.

[19]A. Heller, "Conversion of Hardness Numbers," *American Machinist*, Vol. 70, 1929, p. 525.

it too crowded. Within the range $H_{RC} = 20$ to 50 most authors show differences of 9 to 16 Brinell numbers, but at $H_{RC}=50$ (corresponding with $H_{B10/3000} = 500$) these differences are greater, most likely due to differences in steel balls and to the unreliability of the Brinell test for such hardness. More astonishing is the increased difference in Brinell hardness mentioned by different authors when H_{RC} is smaller than 20. It is conceded that the Rockwell test with a diamond has no practical value for such soft materials. This explains

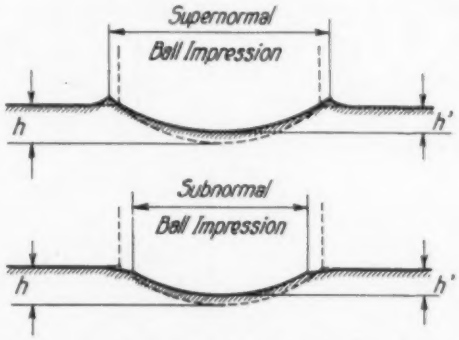

Fig. 5—Monotron Spherical Impression.

why neither Wilson-Maeulen nor any of the authors state the corresponding values within this range. Our tests above $H_{RC} = 30$ agree very closely with tests of all other authors and best with the tests of A. Heller. In lower hardnesses than $H_{RC} = 30$ we have obtained results slightly deviating, that is, relatively low H_{RC} numbers corresponding with Brinell numbers of undoubted accuracy, the latter well in agreement with H_{V50} numbers. This could be explained by the fact that the diamond was not ball-shaped, that the conversion curve is too steep in that part, so that a small variation in Brinell hardness, coupled with lack of homogeneity, produces great differences in Rockwell hardness. Just the opposite happens with great hardness numbers where the curve is flat and, consequently, the hardness numbers of hardened steel are too close together. This explains also the small dispersion of hardness numbers on this machine in exceptionally hard hardened steels, which we have mentioned before.

The most comparable numbers are the Vickers and the "C-D"

Monotron numbers. As Fig. 4 shows their relation is linear, because Brinell numbers obtained by means of the Vickers machine are proportional to the load which is really at the base of the $H_{MC\text{-}D}$ numbers. The total depth of penetration of a diamond with a ball of 0.75 millimeter diameter is h = 5/1000 inch = 0.0457

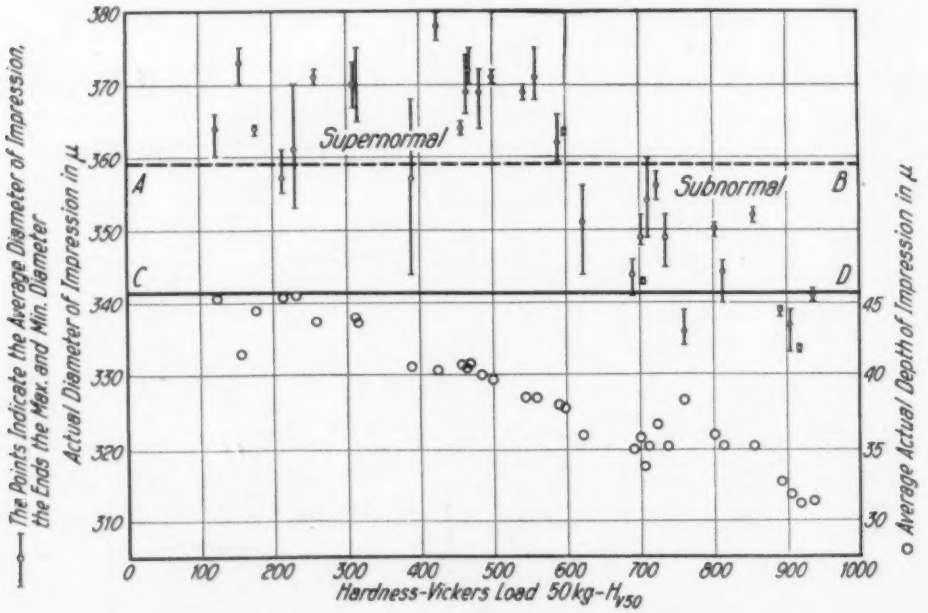

Fig. 6—Actual Diameters and Depths of Impressions on the Monotron.

millimeter. The theoretical surface of the ball-shaped impression is F = π x C.75 x 0.0457 = 0.1076 mm²; the Brinell number on the Monotron called by Shore "Diamond Ball Brinell" would be

theoretically $H'_{MB} = \frac{H_{MC-D}}{F} = \frac{H_{MC-D}}{0.1076} = 9.28H_{MC-D}$. Actually

only plastic deformations are measured while conducting Brinell hardness tests, so that entirely different relations are obtained. By eliminating elastic deformations, that is, by eliminating the load on the Monotron, the penetration of the diamond is reduced to h′ (Fig. 5), which may be easily verified on the depth gage of the instrument. In consequence of the inner structure of the test piece the diameter of the impression will be either larger than normal, i.e., supernormal or smaller, or subnormal (Fig. 5), according to

Shore's glossary[19]. We have measured under a Vickers machine microscope three impressions produced on the Monotron on almost all test pieces.

We have found the impressions to be out of round 10 μ due to the incorrect shape of the diamond. We have, therefore, taken the mean diameters and plotted them in diagram Fig. 6, giving the maximum, minimum and mean values. The theoretical diameter should be 0.359 and is represented by line A-B. Besides that the actual depths of the impressions are shown. Line C-D represents the total depth of penetration of the diamond. The greater the hardness the greater the elastic deformation and the actual depth of the impression will be smaller. However, it is interesting to note that up to $H_{V50} = 600$ the supernormal impression is almost always obtained, while above that figure the subnormal impression predominates. This fact explains why up to $H_{V50} = 600$ the Brinell hardness obtained on the Vickers machine is lower than the theoretical "Diamond-Ball Brinell" hardness obtained on the Monotron $H'_{MB} = 9.28\ M_{MC-D}$, as shown on diagram Fig. 4, by line A-B. The supernormal impression yields a smaller Brinell hardness number than the theoretical value. However, above $H_{V50} = 600$ the subnormal impression on the Monotron predominates and the H_{V50} hardness is greater than the theoretical value. On that account the conversion curve of H_{MC-D} into H_{V50}, shown on diagram Fig. 4 by line A′-B′ follows the equation $H_{V50} = 9.6H_{MC-D} = 20$. Only when $H_{V50} = 580$, $H_{V50} = 9.28H_{MC-D}$. If, for instance, from the Monotron penetration depth the Brinell hardness H_{MB} is figured, we obtain numbers close to H_{V50}. This is recognized by the fact that the hardness numbers H_{MB} are close to the line C-D which connects equal Brinell numbers on both the Vickers machine and the Monotron. From the foregoing it is obvious that the principal reason for the difference between H_{V50} and $H'_{MB} = 9.28H_{MC-D}$ is based on the fact that Monotron tests measure the plastic and elastic deformations of

[19] A. F. Shore, "The Monotron Hardness Indicator," Bulletin M-5, April, 1929, p. 10.

the test piece as total depth of impression. However, our tests have ascertained with a certain amount of probability a straight-line relation of the hardness numbers of both methods, and of the Brinell method up to $H_{V50} = 500$. Therefore, it may be concluded that both methods are producing results which well compare with each other especially when testing hardened steel. The Rockwell method results do not compare very well with both of them when testing hardened steels.

Conclusions

1. The Vickers diamond hardness testing machine is well adapted for individual and laboratory tests of hardened steel especially if the shape of the work is complicated.

2. The Rockwell machine has its place when work of simple shape has to be tested on a production basis.

3. The Monotron machine may be of advantage when tests on a production basis have to be performed and at the same time the state of the surface has to be investigated.

4. Brinell hardness numbers $H_{B\ 10/3000}$ up to 500 and Monotron hardness numbers $H_{MC\text{-}D}$ show in comparison with Vickers hardness numbers a linear relation.

5. C-Rockwell numbers from H_{RC} 20 to 50 may be well converted into Brinell, Vickers and Monotron hardness numbers. Outside of this range the conversion is not accurately possible, as the relation between the Rockwell numbers and the others deviates from a straight line.

THERMOMAGNETIC PHENOMENA IN STEEL AND THEIR APPLICATION IN THE STUDY OF TEMPERING OF QUENCHED 0.75-PER CENT CARBON STEEL

I. Fundamental Considerations

By Raymond L. Sanford and George A. Ellinger

Abstract

The paper gives a brief outline of the principal effects of heating and cooling on the magnetic properties of steel. It is pointed out that the magnetic transformations are associated with the well known thermal transformations. The observation of magnetic changes which occur as the result of heating or cooling and their interpretation in terms of the corresponding structural transformations has been called thermomagnetic analysis. The method is very sensitive, and by its use changes in condition or structure may be observed which apparently can be observed in no other way.

An application of the method to the study of the decomposition of the constituents of water-quenched 0.75-per cent carbon steel has been made. Immediately upon the application of heat, an increase in magnetization took place, which is believed to represent stress release, caused perhaps by the precipitation of carbon from the supersaturated solid solution of carbon in alpha iron. The degree of tempering is believed to be dependent upon temperature and time, the first stage being completed at 200 degrees Cent. (390 degrees Fahr.) in about six and one-half hours. The "precipitated carbon" is believed to combine with iron in the range between 200 and 300 degrees Cent. (390-570 degrees Fahr.) to form iron carbide. This combination is progressive with time, being completed at 300 degrees Cent. (570 degrees Fahr.) in about one hour. From 300 to 700 degrees Cent. (570-1290 degrees Fahr.) no further magnetic changes were noted upon holding at constant temperatures, consequently

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A paper presented before the Thirteenth Annual Convention of the society held in Boston, September 21 to 25, 1931. Raymond L. Sanford is chief of magnetic measurements section and George A. Ellinger is assistant metallurgist, Bureau of Standards, Washington, D. C.

no further constitutional changes were believed to take place in this range, but the carbide apparently coalesced to form larger particles. An increase in magnetization at about 235 degrees Cent. (455 degrees Fahr.) was interpreted as indicating the decomposition of austenite, the decomposition being completed in about 45 minutes at that temperature.

Introduction

SINCE Gilbert, in 1600, discovered that red-hot iron is not attracted by a magnet, the effect of temperature on magnetism has been the subject of numerous investigations. At first, attention was confined to the simple ferromagnetic metals, iron, nickel and cobalt, and the problem was attacked from a purely physical point of view. In later years, however, investigations have been extended to steels of various compositions, and it has been discovered that observations of the magnetic reactions accompanying heating or cooling offer a promising means for studying the corresponding transformations occurring within the structure of ferromagnetic alloys. For the sake of brevity, such observations with special reference to their interpretation in terms of the corresponding structural transformations have been called thermomagnetic analysis.

The Process of Magnetization

The magnetization of ferromagnetic materials progresses in three more or less distinct stages as shown in Fig. 1. It is thought to consist in the orientation of minute magnetic elements existing within the material. As an aid to the imagination in visualizing the phenomena of magnetization, the process may be thought of as follows:

In the neutral or demagnetized condition, the elementary magnets arranged in definite groups, fixed by the crystal structure and conditions of the material, are oriented in such a way that the lines of magnetic flux form closed figures entirely within the material so that there is no external effect. Upon the application of a small magnetizing force, the elementary magnets are turned slightly toward the direction of the applied force, but the forces of attraction tending to maintain the original configuration are relatively large and the magnetization increases at a comparatively slow rate with increase in magnetizing force. This is the first stage of magnetization. Within

this stage, the effect of previous magnetic history is most conspicuous, and, therefore, observations taken in this range must be preceded by thorough demagnetization.

During the transition from the original or neutral configuration each group of elementary magnets passes through a condition

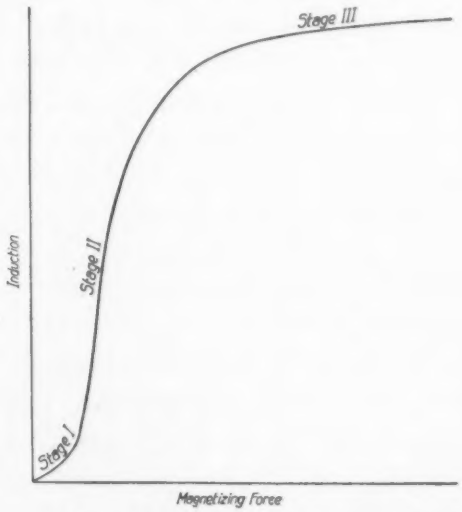

Fig. 1—Normal Induction Curve Showing Three Stages of Magnetization.

of instability and at a certain point suddenly takes up a new and more stable configuration. The range of magnetizing force in which

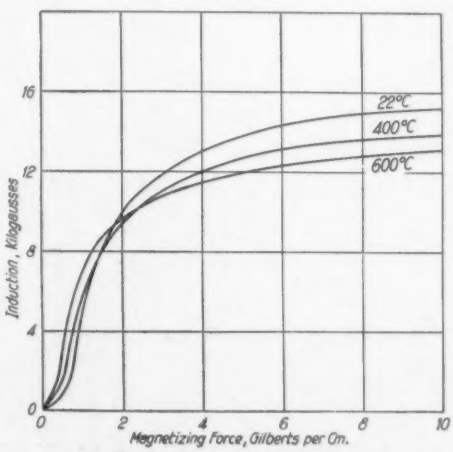

Fig. 2—Normal Induction Curves for Iron Showing the General Effect of Temperature.

the sudden transitions take place is represented by the steep part of the magnetization curve, which constitutes the second stage of

magnetization. In this stage, strain, vibration, temperature, and other temporary influences have their maximum effect.

After all of the magnetic groups have made the sudden change, the third stage of magnetization begins. The mutual forces between the elementary magnets oppose the orienting effect of the applied magnetizing force more and more as parallelism is approached and the slope of the curve becomes less as the magnetizing force is increased. In the third stage, the magnetic properties depend more on the actual constitution of the material and less on temporary effects such as strain and previous magnetic history.

General Effect of Temperature on Normal Magnetization

The general effect of temperature on the normal induction of annealed pure iron is shown in Fig. 2. In the lower part of the curve, heating above room temperature causes an increase in permeability. At higher magnetizing forces the permeability is decreased, the reversal in effect coming at different values of magnetizing force as the temperature is increased. At low values of magnetizing force, the magnetization corresponding to a given value of magnetizing force

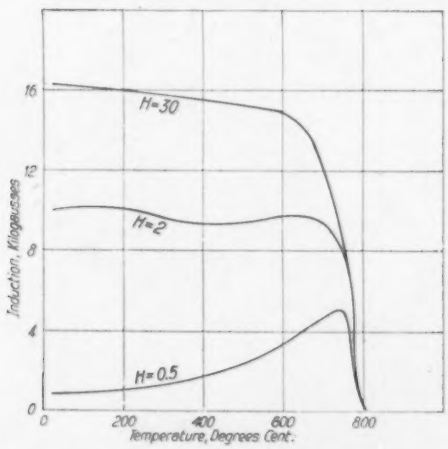

Fig. 3—Effect of Temperature on Induction for Different Values of Magnetizing Force.

(Fig. 3) continually increases with increase in temperature until a maximum is reached at about 750 degrees Cent. (1380 degrees Fahr.), after which it suddenly drops to practically zero at about 810 degrees Cent. (1490 degrees Fahr.). As the magnetizing force is increased, this peak in the curve gradually diminishes until, at a suffi-

ciently high value, the magnetization decreases continually from the beginning of heating to the critical temperature. This critical temperature at which ferromagnetic properties completely disappear is called the Curie point, after the French physicist who made many important investigations of this phenomenon. The way in which the Curie point is approached depends upon the value of the magnetizing force, but the temperature is independent of the force.

There are three factors involved in the changes in magnetic properties which occur on heating, (1) thermal agitation, (2) change in the relative positions of the magnetic elements within the material due to thermal expansion or variations in the degree of mechanical strain and (3) changes in the magnetic character of the material due to metallographic transformations. The first two factors are always present, but the third may or may not enter according to the nature and initial condition of the material.

The general influence of thermal agitation is to produce disorganization, but in the earlier stages of magnetization it also tends to aid the orientation of the magnetic elements toward the direction of the applied magnetizing force, thus increasing the magnetization. If this effect is greater than the disorganizing effect tending to decrease magnetization, the result is a net increase. For the higher magnetizing forces, the disorganizing effect predominates.

The changes due to thermal expansion, variations in mechanical strain and the like, are especially prominent in the second stage of magnetization and generally are so great as to mask completely any other effect. For this reason, it is usually preferable to avoid working in this stage.

Thermomagnetic Analysis Methods

The most complete and accurate data on magnetic properties as influenced by temperature are obtained by the ballistic method. In the application of this method, long samples, generally in the form of wire, and surrounded by test coils insulated with heat-resisting materials, may be used. It is also possible to employ ring specimens, in which case the magnetizing winding also must have insulation capable of withstanding high temperatures. If long wire specimens are used, the magnetizing winding may be located outside a water-jacketed furnace. In making ballistic observations, the specimen is brought to a given temperature which is then held constant while the magnetic

observations are taken. The temperature is then changed and another set of observations are taken, and so on. This method yields the maximum amount of magnetic data, but has certain inherent disadvantages. If the chosen temperature lies within a range in which some transformation takes place in the material, then the characteristics of the material will be continually changing during the magnetic test, and the curve obtained will not represent a single condition of the material. Furthermore, details in the progress of the transformation between the temperatures at which observations are taken may be lost entirely. For this reason either alternating current or magnetometer methods are usually preferable. Of these, the magnetometer is the more flexible and probably the better, and the following discussion refers primarily to results obtained by this method.

The most serious objection which might be raised to the use of the magnetometer for thermomagnetic measurements is its sensitiveness to extraneous magnetic influences. It has been demonstrated, however, that this difficulty is not insuperable. By proper design, installation and adjustment of the apparatus, it is possible to reduce the effects of outside influences to such an extent that they can be neglected. A magnetometer for thermomagnetic analysis has been in successful operation at the Bureau for more than two years.

The apparatus used in the experimental work described later was a slightly modified form of that previously described by one of the authors.[1]

In using the magnetometer, it is not possible to employ very long specimens, and consequently the self-demagnetizing effect must be taken into consideration. This may be thought of as the influence of the magnetic poles of the specimen, which is equivalent to a magnetizing force in the opposite direction to that of the solenoid. The magnitude of the demagnetizing effect is a function of the ratio of length to diameter of the specimen and also of the degree of magnetization. The actual effective magnetizing force is not only much smaller than the applied force, but also its value changes as the magnetization changes with temperature. The result is to reduce the apparent intensity of the magnetic changes as the specimen is heated or cooled.

In order to obtain satisfactory results, it is desirable to have the ratio of length to diameter as large as possible, preferably not less

[1]R. L. Sanford, "Apparatus for Thermomagnetic Analysis," Bureau of Standards Journal of Research, Vol. 2, (R. P. 50), p. 659; 1929.

than 15. At the same time the length should not exceed that of the region in which the temperature and magnetizing force are substantially uniform. This depends upon the design and dimensions of the furnace and magnetizing solenoid. In work at the Bureau of Standards it has been found convenient to use a specimen 10 centimeters long and 6 millimeters in diameter which has a dimensional ratio of 16.7.

Thermomagnetic Hysteresis

The phenomenon of thermomagnetic hysteresis is exhibited to the greatest degree in the second stage of magnetization. It is illustrated in Fig. 4. It might appear that some permanent change in the properties of the material had resulted from the cycle of heating and

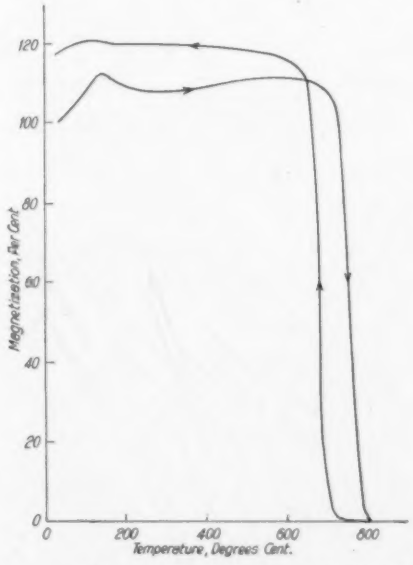

Fig. 4—Thermomagnetic Analysis Curve for Annealed 0.63 per cent Carbon Steel in Second Stage of Magnetization.

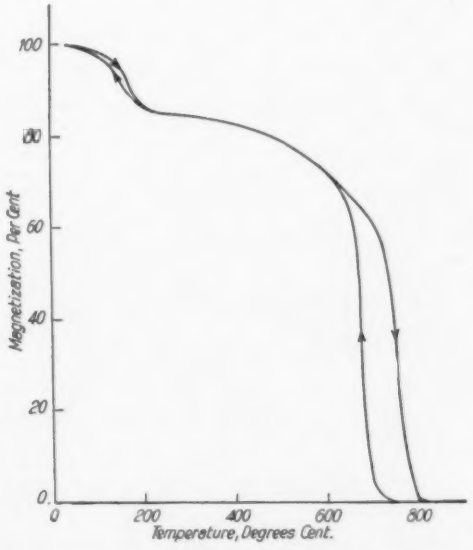

Fig. 5—Thermomagnetic Analysis Curve for Annealed 0.63 per cent Carbon Steel in Third Stage of Magnetization.

cooling to which the specimen was subjected during the test. The material, 0.63 carbon steel, had previously been annealed, however, at 925 degrees Cent. (1700 degrees Fahr.), a hundred degrees higher than the test was carried. Furthermore, when the magnetizing force was reversed several times after the specimen had reached room temperature, the magnetization dropped to the original value. This shows that the difference is associated with magnetic hysteresis rather than with a change in the normal permeability of the material. In this test the effective value of the magnetizing force was about 5 gilberts per

centimeter, giving an induction of approximately 7 kilogausses, which is about the middle of the second stage of magnetization. Fig. 5 shows the result of a test run at an effective magnetizing force of about 150 gilberts per centimeter. The induction was of the order of 16.5 kilogausses, well into the third stage of magnetization. In this case the effect is negligible.

The phenomenon is evidently a magnetic hysteresis effect brought about by the change in temperature and thermomagnetic observations taken in the second stage of magnetization may be difficult to interpret accurately on account of thermomagnetic hysteresis which may easily be confused with the effect of stress release.

Thermomagnetic Transformations

Thermomagnetic transformations are associated with the ordinary thermal transformations and are of two kinds. The changes due to the first kind are such that during the change the magnetiza-

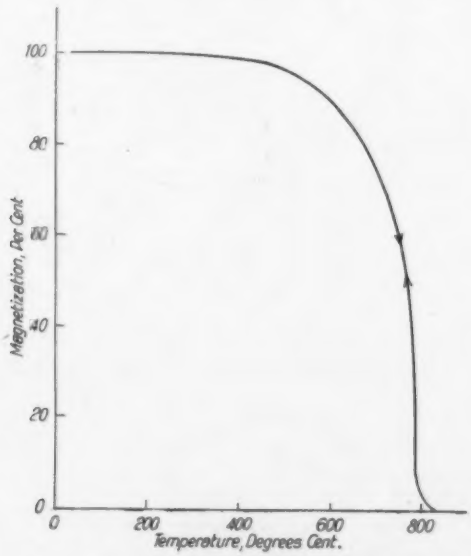

Fig. 6—Thermomagnetic Analysis Curve Showing the A_2 Transformation in Iron.

tion is a function of the temperature alone and does not progress with time, however long the specimen may be held at a given temperature. This is best illustrated by the A_2 transformation shown for electrolytic iron in Fig. 6. The nature of the A_2 transformation is not well understood. The fact that the magnetization is a function of the temperature suggests that the magnetic effect is mainly, if not

wholly, the result of thermal agitation. On the other hand, changes of specific heat, accompanied by heat absorption or evolution within the transformation range, imply some more profound structural change. No change in crystal form is detected by X-ray analysis. According to Honda[2] the criterion for an allotropic transformation is that it will complete itself at a definite temperature if given time enough. The A_2 transformation is not of this type, and we may tentatively conclude, therefore, that it does not represent an allotropic change. Akin to the A_2 transformation is the A_0 transformation observed in the range from 100 to 220 degrees Cent. in Fig. 5. This has been rather definitely associated with the iron carbide, Fe_3C and is most marked in annealed steels having a lamellar pearlite structure.

The second kind of magnetic transformation is one which progresses with time at a definite temperature and will complete itself if given sufficient time. Such a transformation meets Honda's criterion for an allotropic change. It would appear, however, that this condition while necessary is not sufficient, as magnetic changes due to release of stress, rejection of carbon from forced solid solution, and combination of iron and carbon to form the iron carbide Fe_3C, all appear to depend upon time as well as temperature, as do the magnetic changes associated with them.

The most conspicuous magnetic transformation of the second kind is associated with the thermal A_1 transformation. In ordinary heating and cooling curves the temperature is progressively changing and the A_1 transformation is indicated by a sudden change in slope of the temperature-magnetization curve which becomes very steep. This corresponds to the transformation of the alpha iron in the pearlite to gamma iron. As can be seen from Fig. 5, the A_1 transformation takes place within the range of the A_2 transformation, which completes itself after the A_1 transformation is over. The intensity of the magnetic change at the A_1 temperature depends upon the carbon content and the temperature is higher on heating than on cooling.

The temperature of the transformation and the magnitude of the lag between heating and cooling may be greatly affected by alloying elements. Nickel, for instance, exerts a large effect.

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[2]K. Honda, "A Criterion for Allotropic Transformations of Iron at High Temperature." *Science Reports*, Tohoku University, Sendai, Japan, Vol. 6, 1917, p. 213.

II. Thermomagnetic Analysis Applied to the Study of Tempered Steel

By G. A. Ellinger

Introduction

The decomposition or tempering of martensite is of great importance to the metallurgist interested in the heat treatment of steel. Thermomagnetic analysis reveals, in terms of changes in magnetic properties upon heating or cooling, certain constitutional transformations which are difficult if not impossible to follow by means of most of the other methods used for the purpose. It was thought, therefore, that it might be possible to obtain by this method a fairly complete picture of the structural changes, taking place upon the tempering of quenched steels.

The condition of the carbon in the quenched state and that occurring as the tempering process proceeds is of considerable importance and is not definitely known. Any light that can be thrown on this question, therefore, should aid materially in the solution of this problem. It has been very well established by several investigators that cementite undergoes a magnetic transformation, similar to the A_2 transformation in iron, at a temperature of approximately 215 degrees Cent. (420 degrees Fahr.). This transformation, known as the A_0 transformation, is clearly shown by curves obtained by thermomagnetic analysis and therefore, serves as a valuable criterion in the interpretation of results. It is of utmost importance to distinguish clearly between variations in magnetization which are due to actual structural changes and those which may be of purely magnetic origin.

Honda[3] has concluded that cementite is in solid solution in quenched steels and that upon gradual rise of tempering temperature to 400 degrees Cent. (750 degrees Fahr.), the carbide is precipitated.

Ishiwara[4] investigated the tempering of 0.94 per cent carbon steel and found an increase in magnetization at 130 degrees Cent. (265 degrees Fahr.), which he attributed to the tempering of martensite to troostite.

[3]Kotaro Honda, "On the Magnetic Investigation of the States of Cementite in Annealed and Quenched Carbon Steels." *Science Reports,* Tohoku Imperial University, Sendai, Vol. 6, p. 149, 1917.

[4]Torajiro Ishiwara, "On the Magnetic Analysis of Carbides Found in Different Kinds of Steel." *Science Reports,* Tohoku Imperial University, Sendai, Vol. 6, p. 285, 1917.

Saito[5] investigated quenched and tempered steels and concluded that upon tempering a quenched steel at 300 degrees Cent. (570 degrees Fahr.), cementite was first set free but owing to the extremely fine particle size, it decomposed rapidly into iron and carbon, which subsequently recombined again to form cementite.

Heyn and Bauer[6] determined the solubility in acids, of steels treated in various ways. Quenched steel, tempered at 400 degrees Cent. (750 degrees Fahr.), was the most soluble, the rate of solubility decreasing in either direction, toward martensite on one side and pearlite on the other. Troostitic specimens left no residue of carbide, but a residue of carbon containing little or no iron. The authors suggested that some substance which they called "X" separated from martensite upon tempering up to 400 degrees Cent. (750 degrees Fahr.). This substance differed greatly from cementite in its reactions with hydrochloric and sulphuric acids.

Maurer[7] investigated the magnetic properties of various steels and found results similar to those of Heyn and Bauer.

Dearden and Benedicks,[8] investigated various annealed and quenched steels by means of thermomagnetic analysis. They found a slight increase in magnetization, for annealed steels, at about 130 degrees Cent. (265 degrees Fahr.) and a point of decreased magnetization, at 210 degrees Cent. (410 degrees Fahr.). Steels quenched from 910 degrees Cent. (1670 degrees Fahr.) showed weak changes in magnetization at 120 and 260 degrees Cent. (248-500 degrees Fahr.).

Duell[9] investigated the magnetic properties of carbon steels by means of a ballistic galvanometer method. Upon heating quenched specimens to high temperature, he found a discontinuous drop in magnetism at 300 degrees Cent. (570 degrees Fahr.), explaining this as "possibly being due to the formation of iron carbide, the carbon of which was finely divided throughout the mass." Duell found no evidence which would indicate the formation of troostite from austen-

[5]Seizo Saito, "On the State of Carbon Steels, Quenched and Tempered." *Science Reports*, Tohoku Imperial University, Sendai, Vol. 9, p. 281, 1920.

[6]E. Heyn and O. Bauer, "Influence of the Treatment on the Solubility of Steels in Sulphuric Acid," *Journal*, Iron and Steel Inst., Vol. 79, p. 109, 1909.

[7]E. Maurer, "Recherches Sur la Trempe et le Revenu du Fer et de L'acier," *Revue de Metallurgie*, Vol. 5, p. 711, 1908.

[8]W. H. Dearden and C. Benedicks, "Magnetic Changes in Iron and Steel Below 400 Degrees Cent." *Journal*, Iron and Steel Inst., Vol 113, p. 393, 1926.

[9]Clifford C. Duell, "Magnetic Investigations of Carbon Steel," TRANSACTIONS, American Society for Steel Treating, Vol. 15, p. 630, 1929.

ite, and concluded that austenite tempers at a "lower temperature and is probably not a very abrupt change." In this investigation a rather low magnetizing force of the order of 7.6 gilberts per centimeter was used.

Enlund[10] made electrical resistance measurements on eight carbon steels, and found that martensite started to temper to troostite at 100 degrees Cent. and that the transformation was most marked at 110 degrees Cent., while austenite tempered at about 260 degrees Cent. (500 degrees Fahr.) to troostite. Results reported recently by Sefing[11] on the energy evolved during solution of quenched and tempered steel, obtained by means of an adiabatic calorimeter are in accord with the results of the present investigation. The maximum evolution of heat was obtained with specimens tempered at 150 degrees Cent. (300 degrees Fahr.) and a second maximum was obtained with specimens previously tempered at 350 degrees Cent. (660 degrees Fahr.).

General Considerations

As previously pointed out it is of great importance, in the interpretation of the results of thermomagnetic analysis, to distinguish clearly between changes due to constitutional transformations and those due to normal magnetic origin.

For the type of steel under investigation it is generally necessary to apply a magnetizing force of at least 100 gilberts per centimeter to bring the material into the third stage of magnetization, in which it is preferable to work. In the present investigation, the value of the applied magnetizing force was always 277 gilberts per centimeter. For the dimensional ratio of the specimens used, the demagnetizing factor was of such a value that the effective magnetizing force was never less than 150 gilberts per centimeter.

The inflection in the magnetization-temperature curve in the neighborhood of 200 degrees Cent. has been the subject of study by various investigators, notably by Honda and his associates,[12] [13] and

[10]B. D. Enlund, "On the Structure of Quenched Carbon Steels," *Journal*, Iron and Steel Institute, Vol. 111, p. 305, 1925.

[11]Frederick G. Sefing, "Thermodynamics of a Heat Treated Steel," Bulletin, Michigan Engineering Experiment Station, East Lansing, Michigan, Vol. 6, No. 7, May 1931.

[12]Kotaro Honda and Hiromu Takagi, "On the Magnetic Transformation of Cementite." *Science Reports*, Tohoku Imperial University, Sendai, Vol. 4, 1915, p. 161.

[13]Kotaro Honda and Takejiro Murakami, "On the Thermomagnetic Properties of the Carbides Found in Steels." *Science Reports*, Tohoku Imperial University, Sendai, Vol. 6, 1917, p. 23.

is generally agreed to be due to the magnetic transformation of cementite, Fe_3C. The temperature at which this change is complete on heating and begins on cooling has been established by Honda, as 215 degrees Cent. (420 degrees Fahr.). The A_0 transformation in pearlitic steels increases in intensity as the carbon content increases. The minimum carbon content, at which the transformation will be found, is not definitely known. However, from unpublished records of the Magnetic Section of the National Bureau of Standards, it has been found that annealed steel containing 0.14 per cent carbon shows a very definite A_0 transformation, the intensity of which is approximately 3 per cent.

It is conceivable that, when cementite is present as one of the constituents of steel, there might be conditions, such as the form and distribution of the carbide, or strain effects, etc., which would modify the transformation temperature just as other transformation points are altered by various alloying elements. There is some evidence in the results here reported that this is the case.

Material

The material studied was a basic open-hearth steel of the following composition:

	Per Cent
Carbon	0.75
Manganese	0.52
Phosphorus	0.018
Sulphur	0.033
Silicon	0.12
Chromium	0.08
Nickel	0.05
Copper	0.06

The specimens were 10 centimeters in length and 6 millimeters in diameter having a length to diameter ratio of 16.7, so that the self demagnetizing effect of the specimens was not excessive. The specimens were normalized after machining to relieve any residual strains which may have been present.

The normalized specimens were packed in carburizing boxes containing fine silica sand and charcoal and heated to 825 degrees Cent. (1515 degrees Fahr.) in an automatically controlled electric muffle furnace. After holding at temperature for 30 minutes, certain specimens were hardened by quenching in a large volume of water, at approximately 20 degrees Cent. Other specimens were allowed to cool in the furnace. Quenched specimens were placed in the apparatus

immediately and normal magnetization-temperature curves taken, using a heating and cooling rate of approximately 2 degrees Cent. per minute. From these curves, temperatures were selected at which to temper other freshly quenched specimens. These specimens were then placed in the apparatus and heated to the various selected temperatures, and held until it was certain that all magnetic changes had been completed. The specimens were then cooled slowly to room temperature, observations being made during both heating and cooling.

Observations and Results

Magnetization-temperature curves for electrolytic iron and for annealed and quenched 0.75 per cent carbon steel are shown in Fig. 7. Magnetization is plotted in per cent, 100 per cent representing the magnetization at room temperature at the start of a run.

Electrolytic iron showed no A_0 transformation, the magnetization remaining constant over the range of temperatures investigated.

The annealed steel showed a very definite pearlitic A_0 transformation ending at about 215 degrees Cent. (420 degrees Fahr.) on heating and commencing at the same temperature on cooling. There was some hysteresis, probably thermal, as is indicated by the difference in percentage magnetization between the heating and cooling curves at 400 degrees Cent. (750 degrees Fahr.). This steel was heated above the critical range and cooled, only those portions of the curves which lay between 0 and 400 degrees Cent. are shown, this being the range of temperatures in which the primary interest of the investigation lay. It was found that between 400 and 700 degrees Cent. (750-1290 degrees Fahr.), no magnetization change with time took place upon holding at various temperatures in this range.

The magnetization-temperature curves, as determined on a specimen freshly quenched in cold water from 825 degrees Cent. (1515 degrees Fahr.), are also shown in Fig. 7. There was a small increase in magnetization up to approximately 125 degrees Cent. (255 degrees Fahr.), after which the increase was more rapid as heating progressed. At approximately 250 degrees Cent. (480 degrees Fahr.), another increase in rate took place, the curve reaching a maximum at about 290 degrees Cent. (555 degrees Fahr.), after which the magnetization rapidly decreased to 400 degrees Cent. (750 degrees Fahr.). Upon cooling, the A_0 transformation, which was entirely lacking on heating, appeared. As in the case of the annealed steel, this specimen

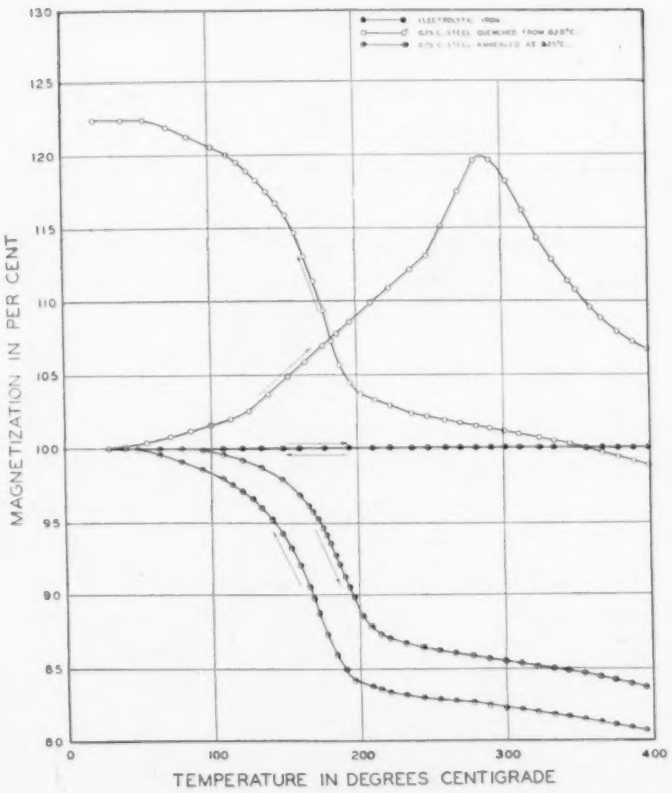

Fig. 7—Magnetization-Temperature Curves for Annealed Electrolytic Iron, Annealed Steel and Quenched Steel Containing 0.75 Per Cent Carbon.

was heated above the critical range and cooled but only those parts of the curves between 0 and 400 degrees Cent. (750 degrees Fahr.) are shown.

The lack of an A_0 transformation could be due to three things: the masking effect of stress release, the presence of carbide in sufficiently finely divided particles so that the transformation might be suppressed, or to the absence of carbide. In order to determine which of these might have been responsible in this case, several investigations were made.

Highly stressed steel shows no A_0 transformation. This is very well shown in cold drawn steel wire. Martensite, as is well known, is in a very highly stressed condition and could mask the A_0 transformation, even if carbide were present.

In order to determine whether the absence of the A_0 transformation was due to the masking effect of stress release, a freshly quenched specimen was heated to 200 degrees Cent. (390 degrees

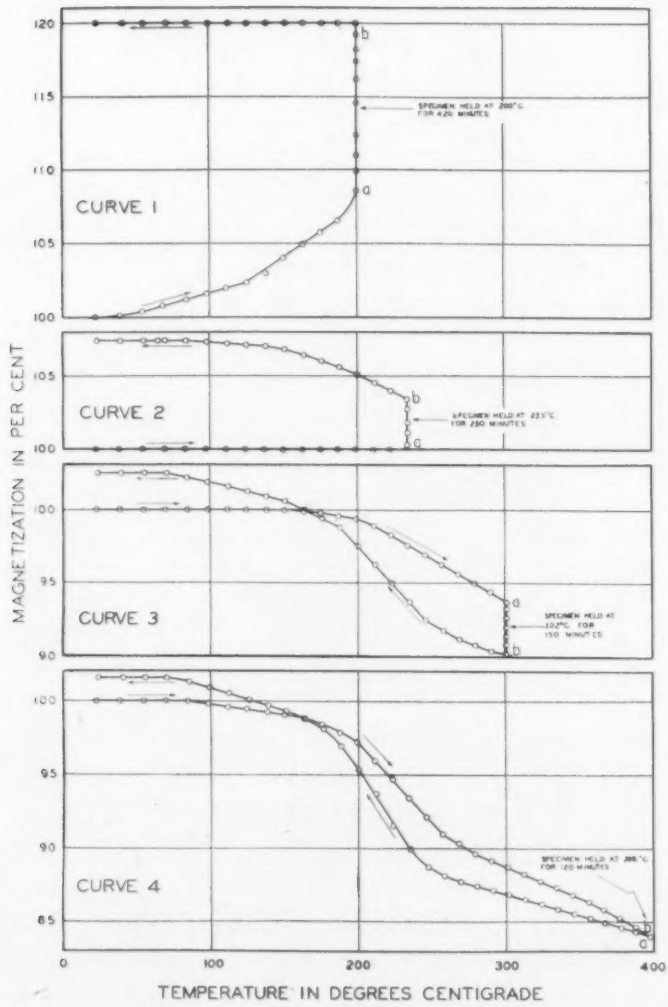

Fig. 8—Magnetization Change with Temperature in Heating a Quenched 0.75 Per Cent Carbon Steel to Various Temperatures. The Portion a-b of each Curve Represents the Change at the Tempering Temperature. Heating and Cooling Curves are Indicated by Directional Arrows.

Fahr.) and held until all change in magnetization was complete. The specimen was then cooled with the results shown in Curve 1, Fig. 8. Curve 1, Fig. 9 shows the change of magnetization with time at constant temperature and represents the portion a-b as indicated on Curve 1 of Fig. 8. Heating and cooling are indicated on the curves by arrows.

At 200 degrees Cent. (390 degrees Fahr.), the magnetization gradually increased for 6½ hours, after which no further change took place at that temperature. There was no change in magnetization upon cooling to room temperature, the curve being a straight line.

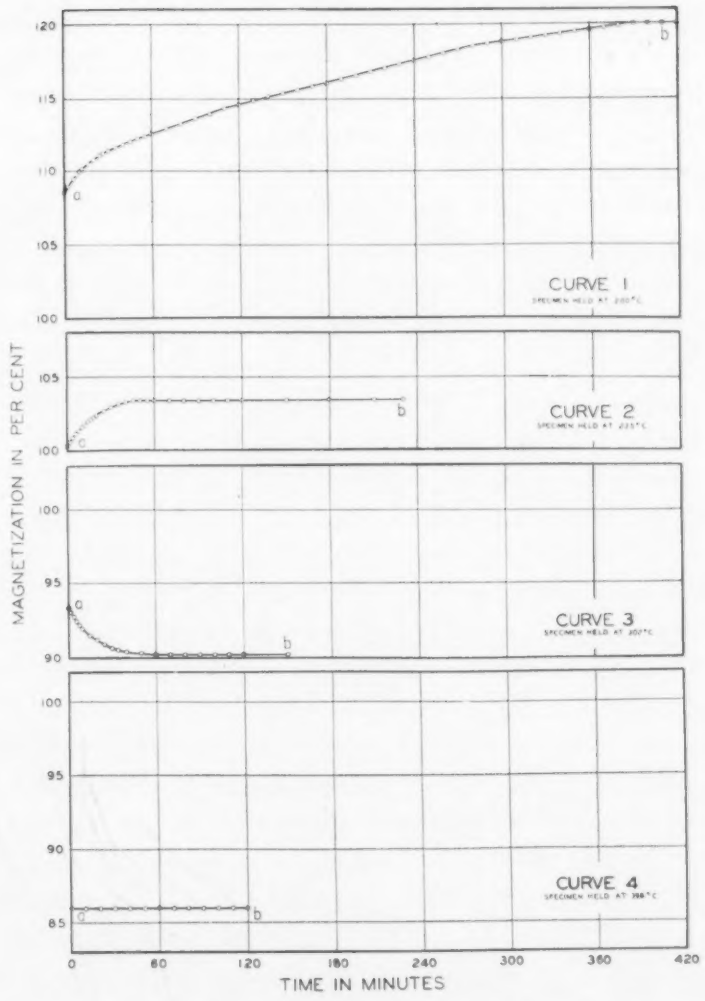

Fig. 9—Magnetization Change upon Holding at Constant Temperatures. These Curves Represent the Portion a-b of the Corresponding Curves Shown in Fig. 8.

Upon reheating, after cooling to room temperature, the magnetization remained constant until a temperature of 235 degrees Cent. (455 degrees Fahr.) was reached, shown in Curve 2, Fig. 8. This indicates that no A_0 transformation had taken place, and that the increase in magnetization shown upon the first heating of the quenched steel to 200 degrees Cent. (390 degrees Fahr.) was probably due to the release of stresses and did not mask an A_0 transformation.

To determine whether or not the presence of carbide in very finely dispersed particles would fail to show an A_0 transformation, a detailed investigation of the effect of carbide size was made. The

results have been reported in another publication.[14] It was found that the intensity of the transformation increased as the particle size decreased. Specimens in which the carbides were so fine as to be barely resolvable under the microscope, had an A_0 transformation, the intensity of which was about half of that of pearlite, whereas specimens containing cementite, spheroidized for various lengths of time, showed a decrease in intensity with growth of particle size.

As a result of these investigations, it is believed that the lack of an A_0 transformation was due to the absence of carbide in the first stages of the tempering of martensite.

Most metallurgists are agreed that martensite is a supersaturated solid solution of carbon in alpha iron. Jeffries,[15] discussing the formation of martensite, said, "The transformation in carbon steel is from a solid solution of carbon in gamma iron to a solid solution of carbon in alpha iron. Only one new crystalline phase is formed, the carbide formation being prevented because the time-temperature conditions do not permit of substantial diffusion of carbon."

Heindlhofer and Bain[16] considered the carbon to be atomically dispersed and the martensite grains to be greatly stressed by their formation. If freshly quenched martensite is considered to contain no carbide and to be in a highly stressed condition, it seems reasonable to attribute the first magnetic change upon heating to 200 degrees Cent. (390 degrees Fahr.) to the release of stress, at least in part. At 200 degrees Cent. (390 degrees Fahr.), the change, as already noted, required over 6 hours for completion. Investigation of tempering at 125 degrees Cent. (255 degrees Fahr.) indicates that the same type of change takes place but that a much greater length of time is required for completion.

Photomicrographs of the material as freshly quenched and after the treatment at 200 degrees Cent. (390 degrees Fahr.), shown in Figs. 10 and 11, give a clear visual indication that the structure was altered considerably by the treatment. The original white martensite needles etched distinctly differently after tempering. The Rockwell hardness also decreased from C 64 to C 56. Cooling Curve 1 and

[14]R. L. Sanford and G. A. Ellinger, "Thermomagnetic Analysis and the A_0 Transformation in 0.75 C Steel," *Proceedings*, American Society for Testing Materials, Vol. 31, Part II, 1931, p. 83.

[15]Zay Jeffries, "A Contribution to the Theory of Hardening and the Constitution of Steel." TRANSACTIONS, American Society for Steel Treating, Vol. 13, p. 369, 1928.

[16]K. Heindlhofer and E. C. Bain, "A Study of the Grain Structure of Martensite." TRANSACTIONS, American Society for Steel Treating, Vol. 18, 1930, p. 70.

heating Curve 2 of Fig. 8 indicate that the change which had taken place in the material at this stage was completed.

Upon reheating to 235 degrees Cent. (455 degrees Fahr.) an increase in magnetization took place, which was completed in about 45 minutes, after which the magnetization remained constant for over three hours. This is shown in Curve 2 of Figs. 8 and 9. This indicated that some constituent, previously not contributing to the magnetization, had become ferromagnetic. Enlund[17] has stated that austenite tempers at 260 degrees Cent. (500 degrees Fahr.). The obvious conclusion is that the retained austenite decomposed at this point, the gamma iron being transformed into the magnetic alpha iron.

Upon cooling to room temperature a slight gradual increase in magnetization was noted. This was presumably the appearance of a carbide transformation.

The microstructure of the specimen at this stage, shown in Fig. 12, indicated that some change had taken place. The structure was very fine and beyond the stage of clear resolution. Some traces of the needles of tempered martensite were still visible. The structure was not that of ordinary troosite as it etched at a somewhat slower rate and had a lighter color than that characteristic of the troostitic range of structures. The hardness was Rockwell C 54, which is nearly the same as after treatment at 200 degrees Cent. (390 degrees Fahr.).

The specimen was then slowly reheated to approximately 300 degrees Cent. (570 degrees Fahr.). This is shown in Curve 3, Fig. 8. A gradual decrease in magnetization took place upon heating, followed by another decrease upon holding at constant temperature. This latter decrease was complete in about 1 hour, no further change being noted upon holding for 1½ hours. The rapid decrease at this stage indicated that a change was taking place within the material, which was effectively decreasing the magnetization of some of the iron.

Upon cooling from 302 degrees Cent. (575 degrees Fahr.), the A_0 transformation was very pronounced as shown in Curve 3, Fig. 8. The temperature at which the A_0 transformation appeared to begin on cooling was approximately 240 degrees Cent. (465 degrees Fahr.) instead of 215 degrees Cent. (420 degrees Fahr.), as observed for this point in the pearlitic structure of the annealed specimen, shown

[17]See foot note 10.

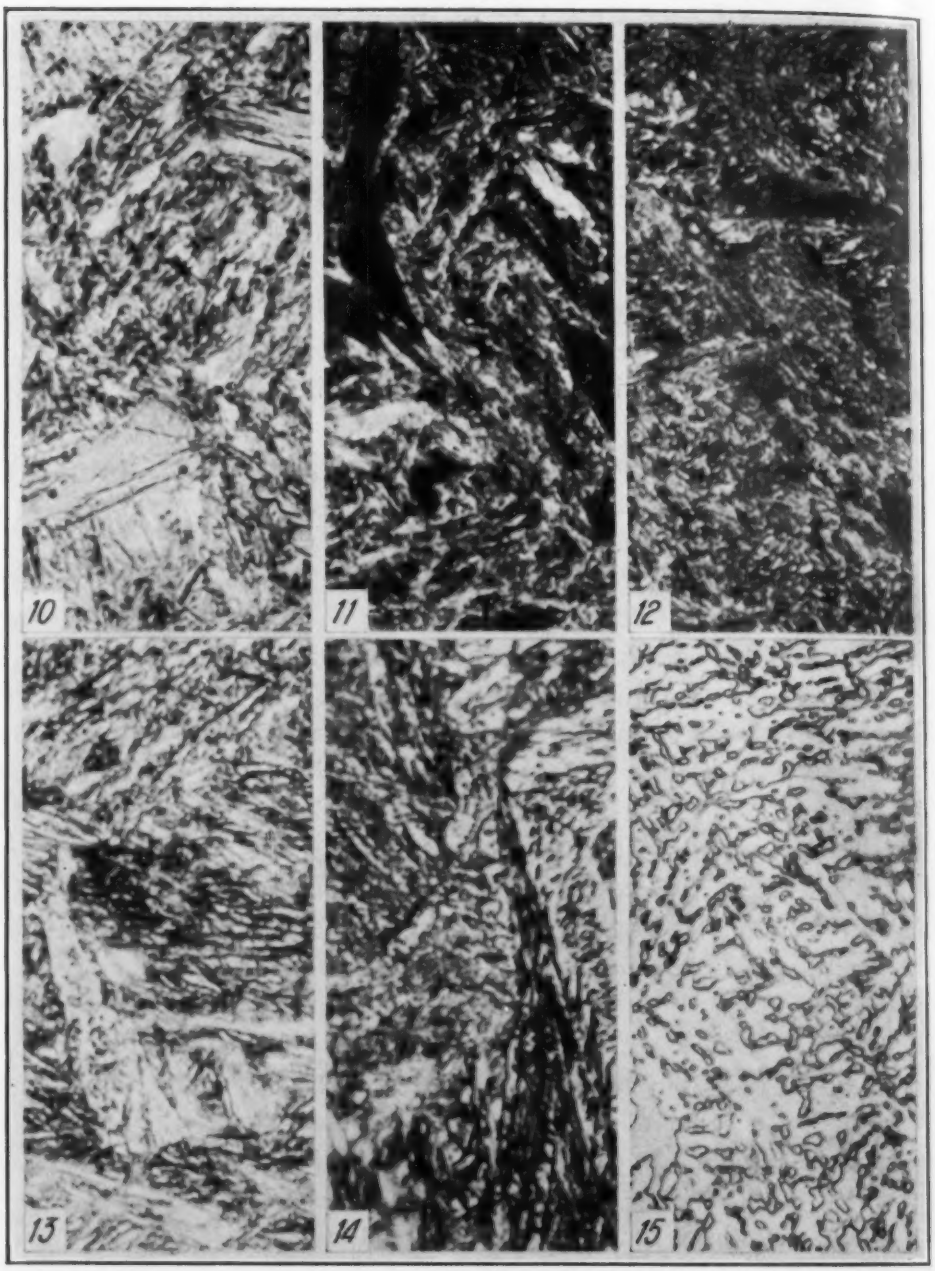

Fig. 10—Structure of Freshly Quenched Specimen. × 2000.
Fig. 11—Structure of Same Specimen as Fig. 10 Tempered at 200 Degrees Cent. for 7 Hours. × 2000.
Fig. 12—Structure of Same Specimen as Fig. 10 Tempered at 235 Degrees Cent. for Nearly 4 Hours. × 2000.
Fig. 13—Structure of Same Specimen as Fig. 10 Tempered at 302 Degrees Cent. for 2½ Hours. × 2000.
Fig. 14—Structure of Same Specimen as Fig. 10 Tempered at 400 Degrees Cent. for 2 Hours. × 2000.
Fig. 15—Structure of Same Specimen as Fig. 10 Tempered at 700 Degrees Cent. for 1 Hour. × 2000.

in Fig. 7. This confirms the results of the previous investigation[18] on the influence of carbide particle size, which showed that the A_0 transformation comes at higher temperatures in material of small particle size.

The structure, shown in D, Fig. 5, was quite different from that of the preceding specimen. The etching rate was somewhat more rapid and the particles could be resolved. The hardness, Rockwell C 49, was somewhat lower than that of the specimen after the 235 degrees Cent. (455 degrees Fahr.) treatment.

When the specimen was heated to 400 degrees Cent. (750 degrees Fahr.) the A_0 transformation was found, but no change in magnetization occurred upon holding at this temperature for 2 hours, as shown in Curve 4, Figs. 8 and 9. Apparently there was no further constitutional change between 300 degrees Cent. (570 degrees Fahr.) and the A_1 transformation, as no magnetic change with time was observed even upon holding the temperature at 700 degrees Cent. (1290 degrees Fahr.) for 1 hour.

The microstructure of the specimen held at 400 degrees Cent. (750 degrees Fahr.), shown in Fig. 14, was not greatly different from that of the specimen held at 300 degrees Cent. (570 degrees Fahr.). The structure was somewhat coarser and the particle size had increased slightly. The Rockwell hardness had decreased to C 40.

The structure of the specimen held at 700 degrees Cent. (1290 degrees Fahr.) shown in Fig. 15, was that of partially spheroidized cementite and the hardness was Rockwell C 12.

All constituents which were stable in the tempering range had been formed at or slightly below 300 degrees Cent. (570 degrees Fahr.), further heating or holding at constant temperatures had apparently no effect in the formation of new constituents as revealed by the magnetic method. Coalescence of carbide particles took place in the range from 300 to 700 degrees Cent. (570-1290 degrees Fahr.) and gave a continuous series of structures ranging from troostite to spheroidized cementite.

Summary

The great sensitiveness of magnetic properties to changes in the condition of ferromagnetic materials affords the basis of a research method by which phenomena associated with the thermal treatment of

[18]See foot note 14.

such materials can be observed, some of which are beyond the scope of other methods. This method has been called thermomagnetic analysis. In order to apply the methods of thermomagnetic analysis properly and to interpret the results with accuracy, a knowledge of the fundamental phenomena is required.

The general effect of heating is to increase the magnetic permeability in the lower part of the magnetization curve and to decrease it in the upper part. The phenomena under relatively low magnetizing forces are complex and difficult to interpret because of the mutually opposing effects of various influences. These confusing effects are much less at higher magnetizations and it is possible to draw valid conclusions from results obtained in the higher range of magnetization.

This method has been utilized in an investigation of the transformations associated with the tempering of a freshly-quenched 0.75 per cent carbon steel and the following phenomena were observed to occur during the process.

1. A pronounced increase in magnetization took place immediately upon the application of heat. After reaching and holding constant at a temperature of 200 degrees Cent. (390 degrees Fahr.), this increase continued for approximately 6½ hours.

2. Another increase in magnetization was noted at about 235 degrees Cent. (455 degrees Fahr.) in a specimen previously treated at 200 degrees Cent. (390 degrees Fahr.).

3. A decrease in magnetization took place upon holding at a temperature of 300 degrees Cent. (570 degrees Fahr.).

4. No change in magnetization with time upon holding at constant temperatures in the range of 400 degrees to 700 degrees Cent. (750-1290 degrees Fahr.) was noted.

The explanation of Jeffries, Archer, Sykes, Bain, and others that martensite is a supersaturated solid solution or pseudo solid solution of atomically dispersed carbon in alpha iron, existing only because the conditions of its formation are such as to prevent carbon precipitation is generally accepted. That martensite is in a highly stressed condition has also been demonstrated by the X-ray investigations of Heindlhofer, Bain and others.

With this in mind, the following hypothetical explanation of the tempering of martensite is suggested as fitting the observed phenomena.

The first stage of tempering freshly quenched martensite below 200 degrees Cent. (390 degrees Fahr.), resulted in an increase in magnetization which was apparently due to stress release. After this change was completed, no A_0 or carbide transformation was found. It is conceivable that, during this stage of tempering, the carbon was being precipitated from the solid solution and remained in dispersed form throughout the ferrite. The absence of an A_0 transformation may be considered to indicate that no carbide had been formed in quantities sufficient to be detected by the magnetic method used.

At approximately 235 degrees Cent. (455 degrees Fahr.), the austenite which had been retained during quenching decomposed and in the range, 200 to 300 degrees Cent. (390-570 degrees Fahr.), carbide formation progressively occurred. After a period of approximately one hour at the latter temperature the formation of carbide was completed. It is possible that this reaction may be completed at a lower temperature, however, in longer periods of time. At higher temperatures, under the critical range, however, no further magnetization changes with time took place, a fact which suggests that no further constitutional changes involving phase changes in the Fe-C system occur in steel under these conditions.

DISCUSSION

YAP, CHU-PHAY:[1] I have not had an opportunity to prepare a written discussion of this paper, but after listening to Mr. Rawdon present this paper I shall not allow the occasion to pass without conveying to the authors through Mr. Rawdon my congratulations. I am glad to know that the Bureau of Standards is interested in such fundamental studies as magnetic investigations of metals and their alloys.

Recently there appeared a paper from Professor Honda's laboratory at Tohoku Imperial University, in which they found the Curie point (A_0 of spheroxidized cementite) to be distinct, but not in the case of troostite, not to mention the more dispersed form in tempered martensite. The work of Messrs. Sanford and Ellinger is, therefore, a good verification. Just why this should be is still an open question, and so the evidence that no Curie point could be found in martensite is hardly acceptable as proof that martensite is a solid solution, even if we ignore the rather conclusive photomicrographic evidence obtained by Dr. Lucas.

In the recent controversy regarding the precipitation of carbide or carbides in the KS-2 type of steels, Mr. Sanford and his associates should be in position to render us considerable help in their studies, because we ought to be able to infer the nature (and, with proper "control," the composition) of the carbide or car-

[1]Consulting metallurgist, 27 Grove St., New York City.

bides present. It is perhaps needless to note that each carbide has a distinct Curie point and even in the case of cementite, the substitution of manganese or chromium in its lattice will be indicated by the displacement in the Curie point.

Another problem of great interest, I think, is the effect of the solutes on the A_2 point in iron alloys. We know, for example, that in the Fe-Zn and Fe-Si system, the A_2 point drops continuously with increasing solute up to the maximum solubility point. This is, however, not the case in iron-carbon alloys, as shown in Fig. 1 below. This point needs to be cleared up because it has some fundamental significance, especially in view of the new theories of ferromagnetism.

It is perhaps not out of place here to mention that Mr. Sanford* has recently shown that a piece of malleable cast iron actually shows the A_0 point after heating it to 800 degrees Cent. (1470 degrees Fahr.) seven times, from

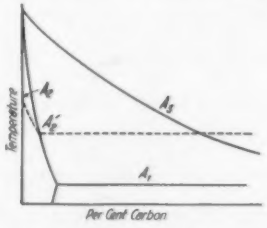

Fig. 1—Note that the A_2 Points Drops to A'_2 (About 0.02 Per Cent Carbon) but the Maximum Solubility is at 0.035 to 0.037 Per Cent Carbon.

which we legitimately conclude that some of the carbon combined to form Fe_3C. I have maintained for several years that Fe_3C must be stable at some high temperature. In our study on the free energy of formation of Fe_3C, we have found that ΔF at 800 degrees Cent. (1470 degrees Fahr.) is -112 cal./mol., increasing as the temperature increases.

E. E. THUM:[2] I would like to point out what seems to be a conflict in the results obtained by Ellinger at the Bureau of Standards from magnetic observations with those obtained by Westgren and others from X-ray work.

Ellinger's curves show that quenched steel has a slight increase in magnetism as the temperature is raised to 200 degrees Cent. When the temperature is held at that point, the magnetism gradually increases and finally reaches a maximum, and on cooling the magnetism retains that maximum. The author ascribes this increase of magnetization to release of stress, on the supposition that martensite is a highly stressed structure, and no phase changes occur below 200 degrees Cent.

On reheating this same sample, no change in magnetization is noted until it reaches 235 degrees Cent., when a very minute increase in magnetization occurred. The specimen was then held at 235 degrees Cent. and the magnetism increased considerably for about an hour but then stayed stationary for several

*Bureau of Standards, Journal of Research, Vol. 2, p. 659, 1929.

[2]Editor of *Metal Progress*, Cleveland.

hours. On cooling this specimen, the magnetism further gradually increased with decreasing temperature.

Even on this second reheating no magnetic change at A_0 was noted, and consequently Ellinger infers that no carbide is yet present. He believes that during the first tempering, the carbon precipitated from solid solution in martensite remains in dispersed form throughout the ferrite. He says: "At approximately 235 degrees Cent. the austenite which had been retained during quenching decomposed, and carbide formation progressed for approximately one hour whereupon the formation of carbide was complete."

A third reheating of this same sample showed stationary magnetism up to nearly 200 degrees Cent. and a considerable linear decrease in magnetism from then on up to 300 degrees Cent. The specimen held at 300 degrees Cent. further decreased in magnetism for about one hour and then remained stationary. On cooling from 300 degrees Cent., a characteristic increase in magnetism at about 250 degrees Cent. was found and this is ascribed to the magnetic change in the cementite particles which then exist in the steel.

This picture of the decomposition of martensite in quenched high carbon steel is different from that which Dr. Westgren gives in an account of recent structural studies on martensite, published in the August 1931 issue of *Metal Progress*. Dr. Westgren's words are as follows:

"A steel of moderate carbon content, say of about eutectoid composition, when normally quenched consists of a slight amount of austenite mixed with a solid solution of carbon in alpha iron, the lattice of which is distorted into a tetragonal shape by the presence of the carbon atoms. When this product is tempered at such a low temperature as 150 degrees Cent., the latter phase is completely transformed into carbon-free alpha iron in which a multitude of extremely fine cementite particles are embedded. This transformation of crystalline structure does not affect the hardness of the steel to any marked degree, and it is thus evident that a steel may be hard even if it contains no supersaturated solid solution."

Dr. Westgren was asked whether his statement that tetragonal martensite gives birth to extremely fine cementite particles when tempered at 150 degrees Cent. has been substantiated by finding interference lines of cementite in photograms of quenched steels tempered at 150 degrees Cent.

His reply follows:

"In your letter you certainly touch upon a somewhat weak point in the X-ray evidence about martensite. The statement that martensite is decomposed into alpha iron and cementite when heated to 150 degrees Cent. is based on the fact that the iron interference lines of this product have exactly the same position as those of pure alpha iron. If the carbon were kept in solid solution after tempering at this temperature, the lines would be displaced.

"Another question is, of course, whether the carbon dispelled out of the martensite is cementite, or some other carbide phase. I must admit that no cementite lines are visible in the photograms until the quenched steel has been heated to about 300 degrees Cent. But this seems quite natural considering that the crystal grains of the phase containing the carbon are extremely small when formed below 300 degrees Cent. Of course, the possibility of this phase being

something else than cementite cannot be considered as excluded even if it seems fairly small.

"Summing up, it is thus quite certain that the martensite of quenched carbon steel after tempering at 150 degrees Cent. does no longer consist of a solid solution but is a mixture of alpha iron and some phase containing carbon, the latter probably being cementite."

FRANCIS M. WALTERS, JR.:[3] We are indebted to Mr. Sanford and Mr. Ellinger for developing thermomagnetic analysis to a high degree of perfection and for giving us an extremely interesting example of its application. I hope that they may devise apparatus which will furnish equally good results in the hands of less gifted experimentalists.

CLOSURE BY DR. H. S. RAWDON:[4] I am not going to attempt to answer the questions raised but there are a couple of points I would like to emphasize in closing.

You will remember from the slides shown that annealed steel consisting of ferrite and cementite exhibits a decrease in magnetization with increase of temperature. There may be two possible reasons for the decrease—the change of the ferrite and change of the cementite. Probably the rapid decrease is to be attributed to change in the cementite, at least this assumption is supported by Honda's work. The point to bear in mind is that the change in magnetization is a decrease.

Now consider what happens in the quenched steel. Upon heating the magnetization begins to increase. Hold the specimen at 200 degrees Cent. and a marked increase results. Bring it back to room temperature and the steel is more magnetic than it was to start with. Whatever the nature of the change that occurs and even if there is a precipitation of cementite starting below 100 degrees Cent. and completing itself, as some believe, below 150 degrees Cent. that precipitation of cementite could not account for this observed increase of magnetism. The formation of cementite ought to result in a decrease of magnetization. If there is any effect of cementite in decreasing the degree of magnetization, it is overshadowed and neutralized by a much larger change, the final result of which is the great increase of magnetization. There would seem to be nothing else to attribute this to but to a stress effect. If the two opposing changes occur simultaneously, the stress effect would be an increase in magnetization and the other would tend to bring it down and the final result would be the increase observed. Unquestionably the stress effect below 200 degrees Cent., must be much greater in its magnetic result than any possible effect of precipitated cementite on the magnetization.

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[3]Carnegie Institute of Technology, Pittsburgh.

[4]Chief, metallurgical department, U. S. Bureau of Standards, Washington, D. C.

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